JULY 1959

JOURNAL OF THE IRON AND STEEL INSTITUTE

4 GROSVENOR GARDENS · LONDON SWI

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VOL. 192 PART 3

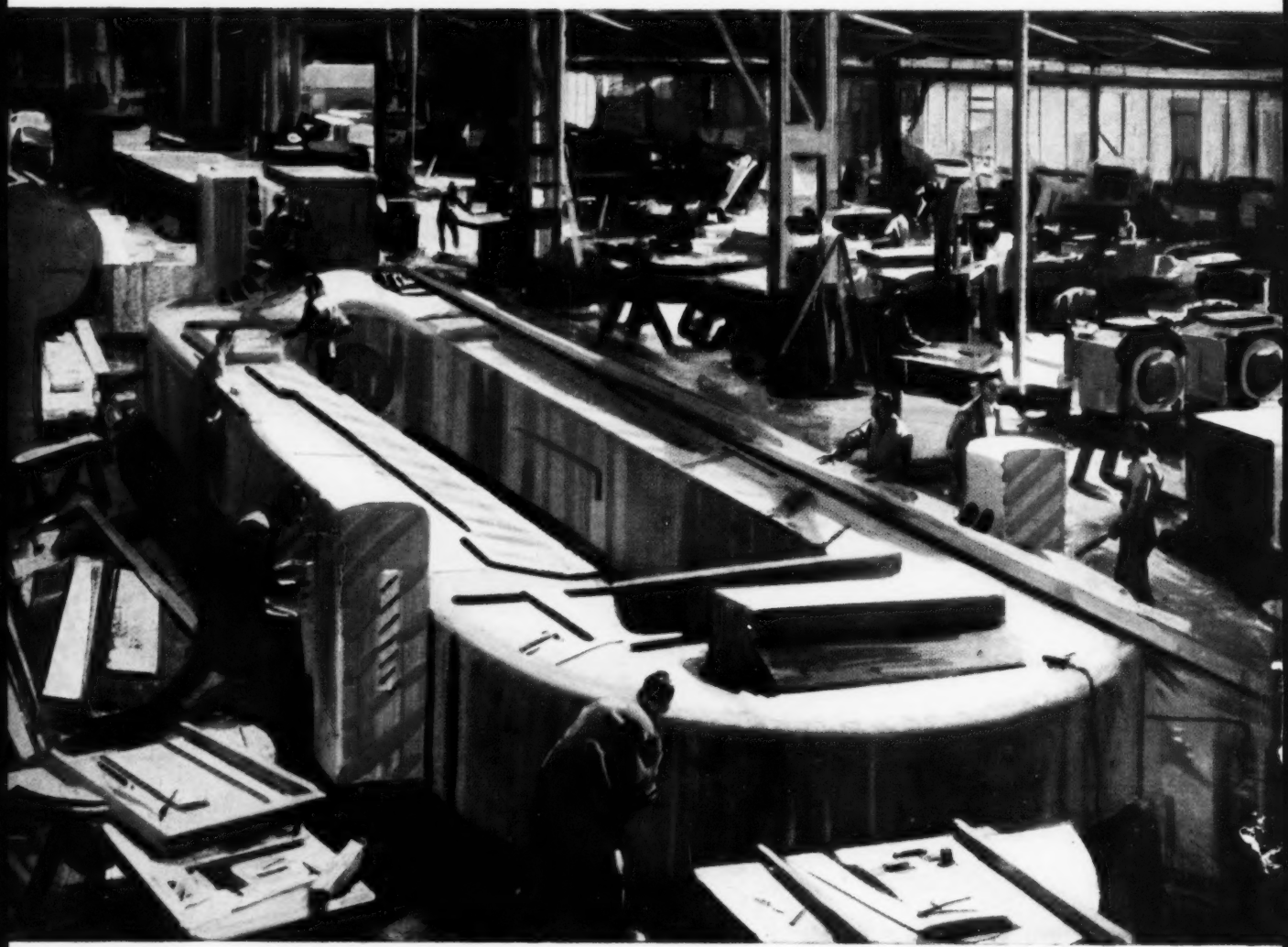
Principal Contents

Review of steelmaking processes *by T. P. Colclough* 201
Effect of copper on martensite breakdown and hardenability of a 0·3% C steel *by A. S. Kenneford* 215
Controlled-transformation stainless steels *by K. J. Irvine et al.* 218
Stress-dependent damping capacity of ferromagnetic metals *by G. Sumner and K. M. Entwistle* 238
Electrolytic jet machining technique for production of thin foils of steel *by P. M. Kelly and J. Nutting* 246
Mechanical properties of hardened high-speed tool steels *by J. Clayton-Cave and E. Ineson* 248
Spectrographic determination of residual elements in steel *by the BISRA Spectrographic Analysis Sub-Committee* 253
Tempering of 2¼% Cr – 1% Mo steel after quenching and normalizing *by R. G. Baker and J. Nutting* 257
Austenite transformation and the Greniger–Troiano technique *by E. G. Ramachandran and C. Dasarathy* 268
Discussion on papers at the Autumn General Meeting, 1958 269
News 298 Abstracts 301

A full list of contents is given on the Title Page (see page 65, Advertisement Section)

Palette on Production

No. 1 in a series of original paintings specially commissioned from Robert Johnston, R.I., S.M.A., illustrating the productive resources for metalworks plant of Davy-United's Darnall Works. The scene shows the making of a pattern for a roll housing for a 12 foot wide hot plate mill for aluminium. When cast the housing weighed 163 tons while the pattern itself took nearly three tons of timber.



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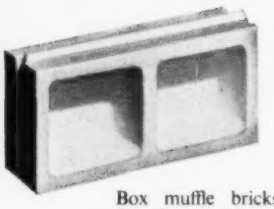
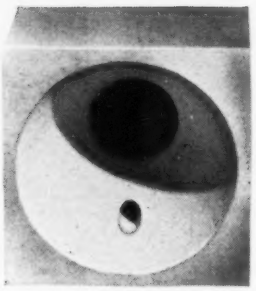
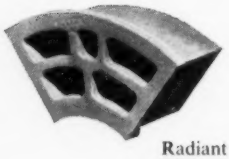


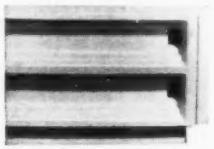
Box muffle bricks for heat-treatment furnaces.



Burner block



Radiant heater block.



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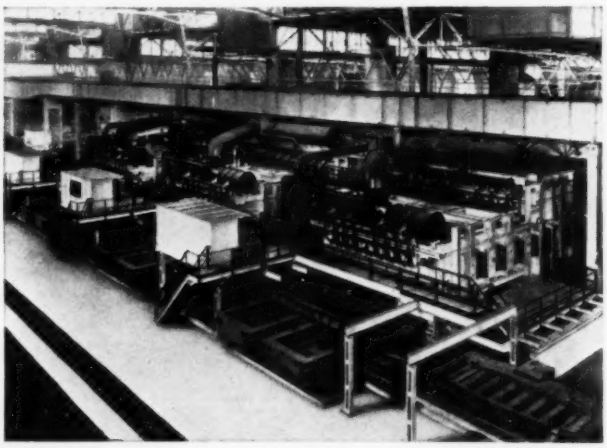

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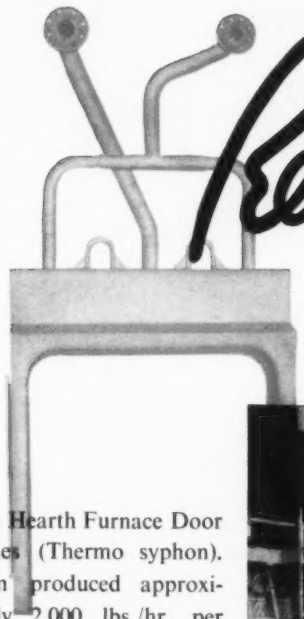
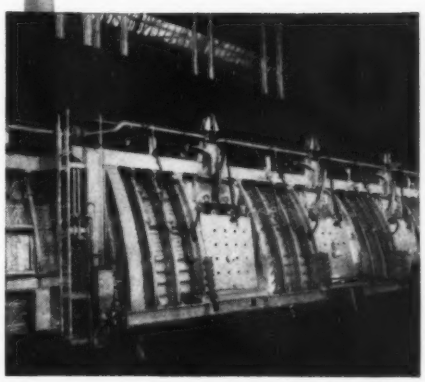


Open Hearth Furnace Door Frames (Thermo syphon). Steam produced approximately 2,000 lbs./hr. per door frame.



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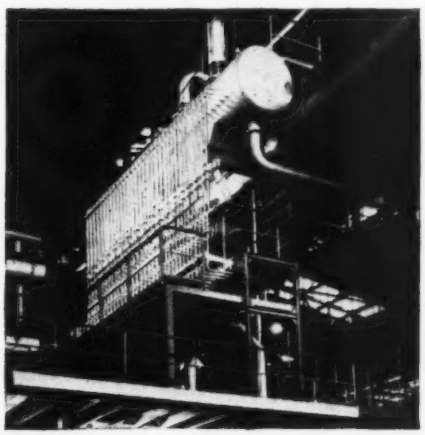

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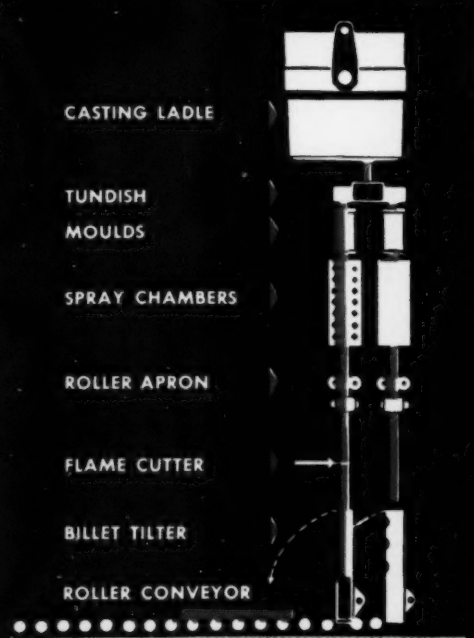

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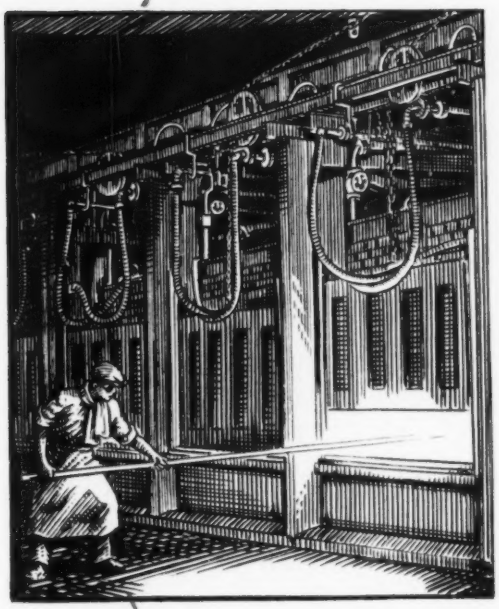

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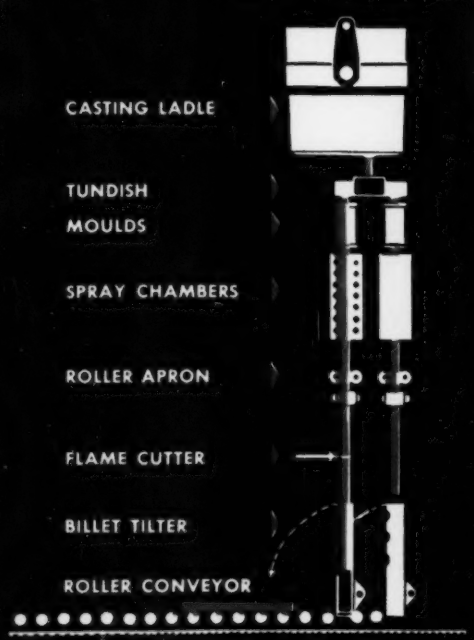

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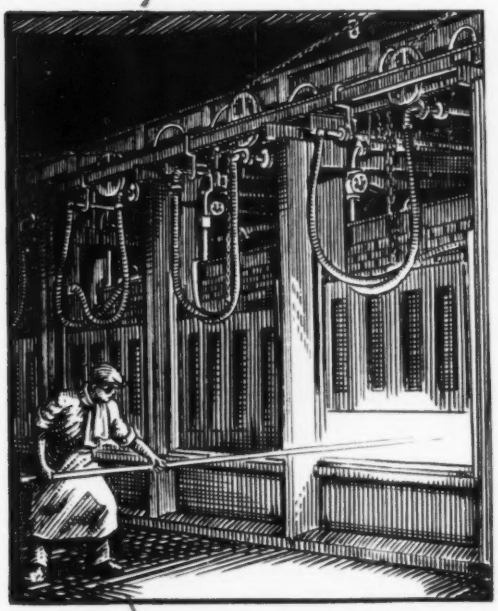

RANGE OF PRODUCTS

Billets from 3" sq. upwards.
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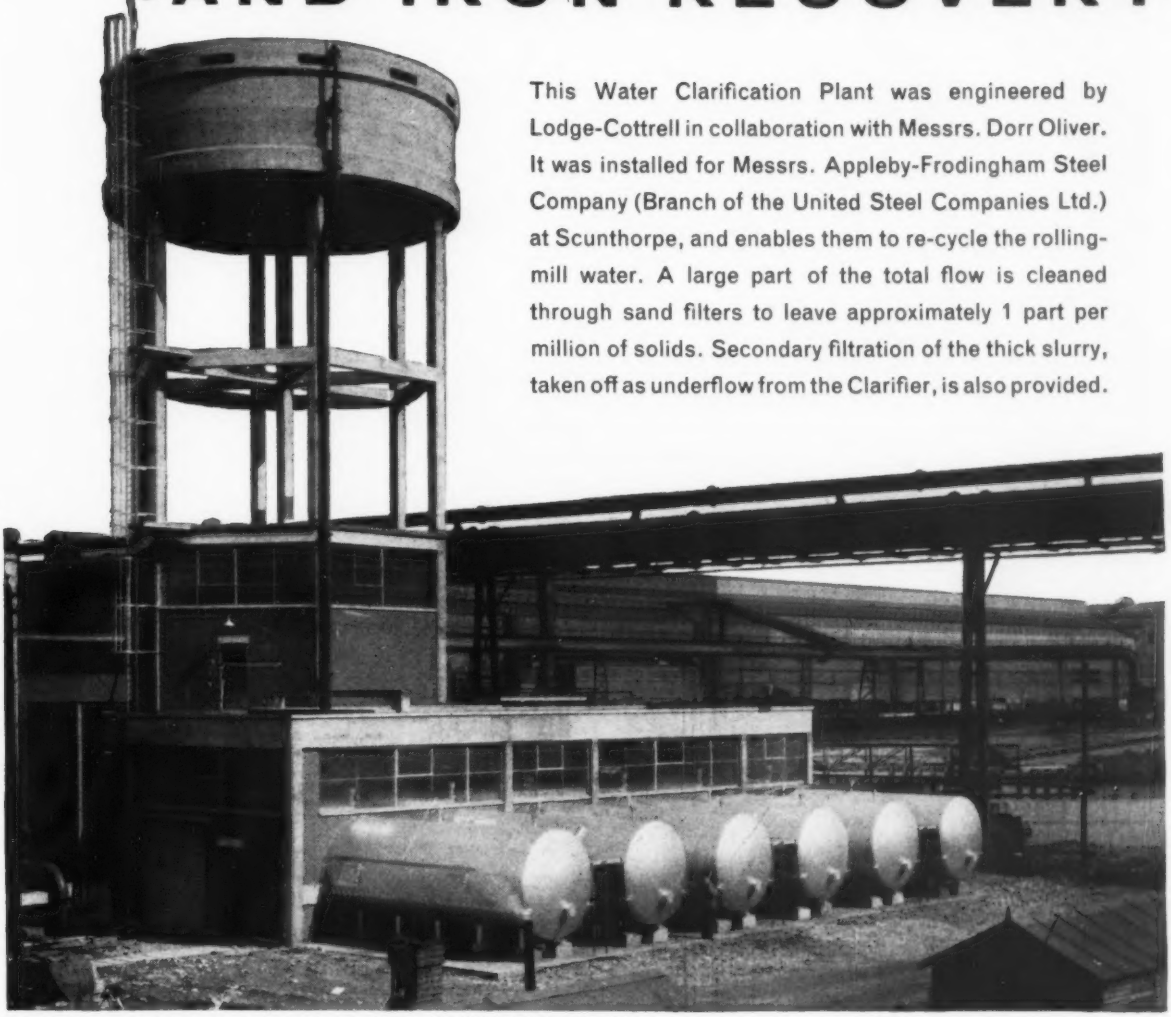

Photograph by courtesy of the Appleby-Frodingham Steel Co. (Branch of the United Steel Companies Ltd).

Another Filtration Plant, similarly engineered, has been in operation for three years at the Ebbw Vale Works of Richard Thomas and Baldwins Ltd. It recovers solids with an iron content of about 35%.

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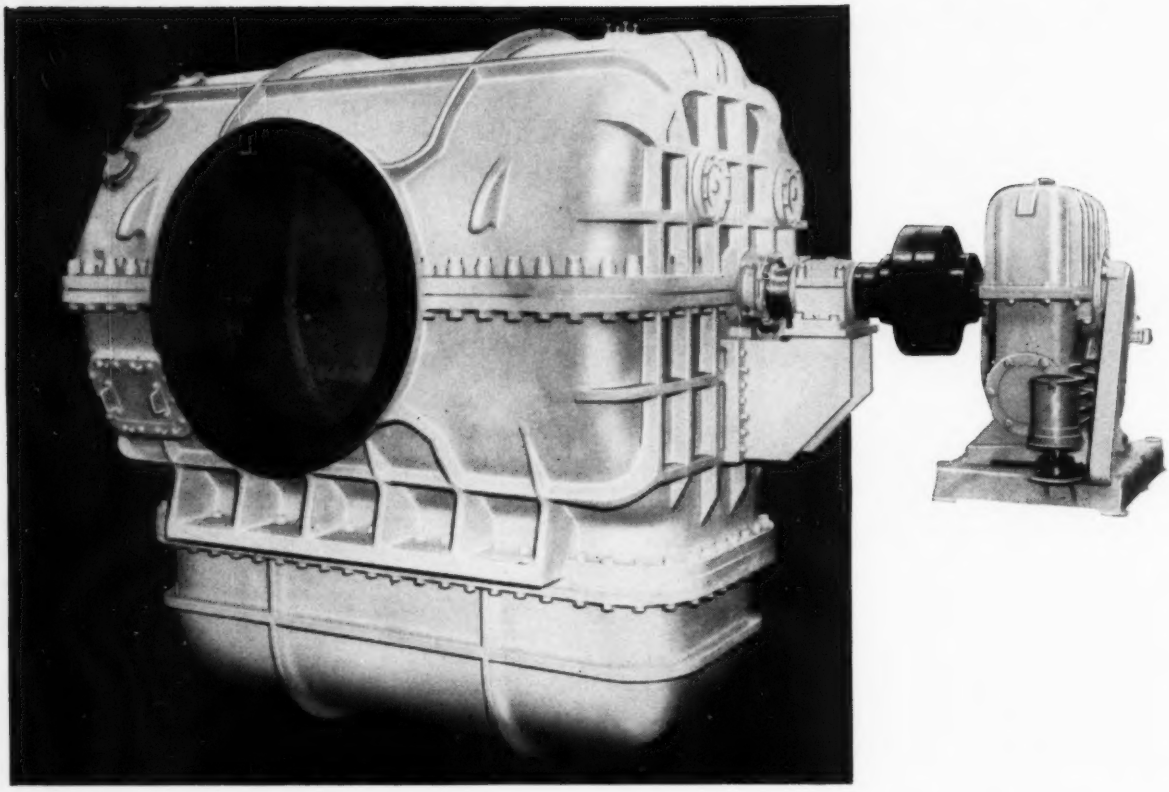

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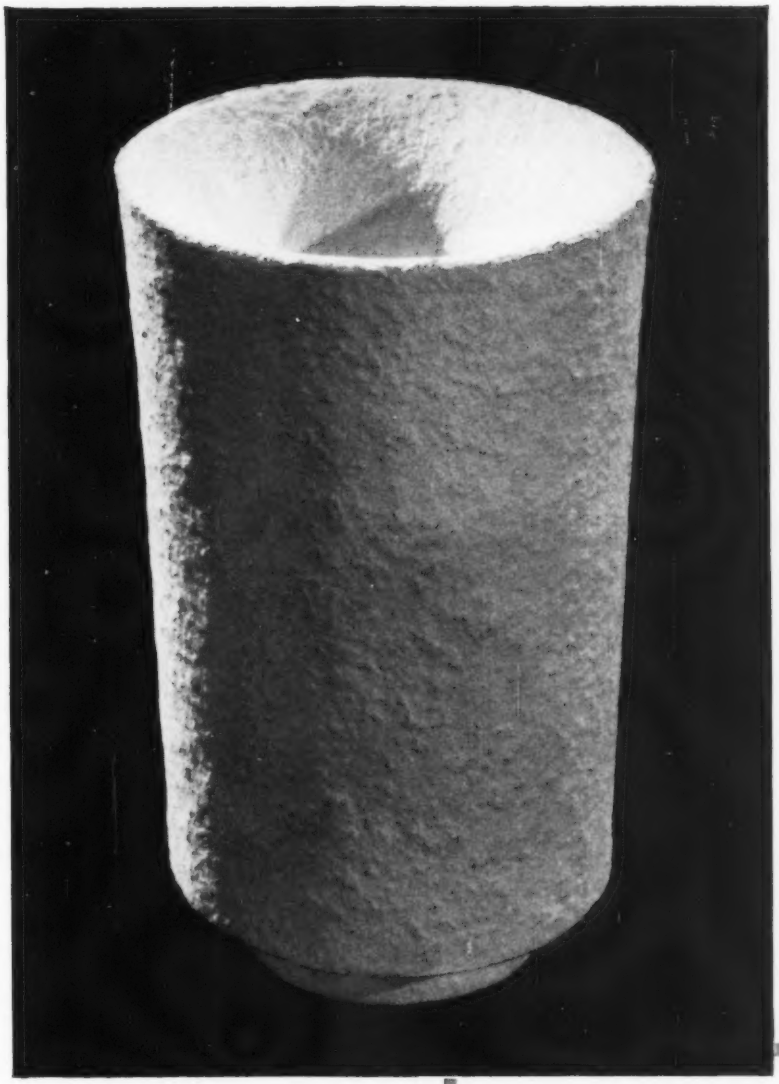

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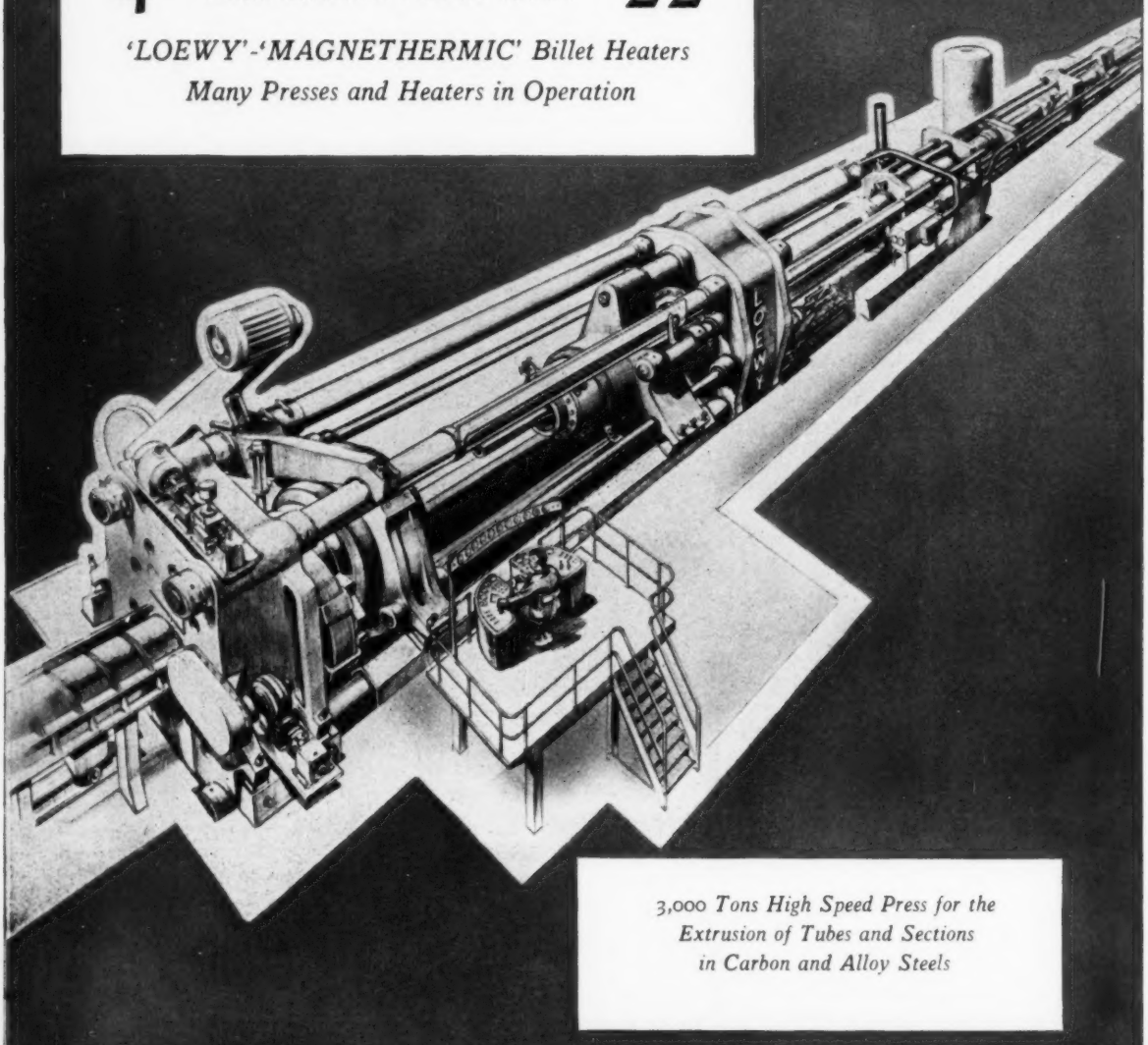

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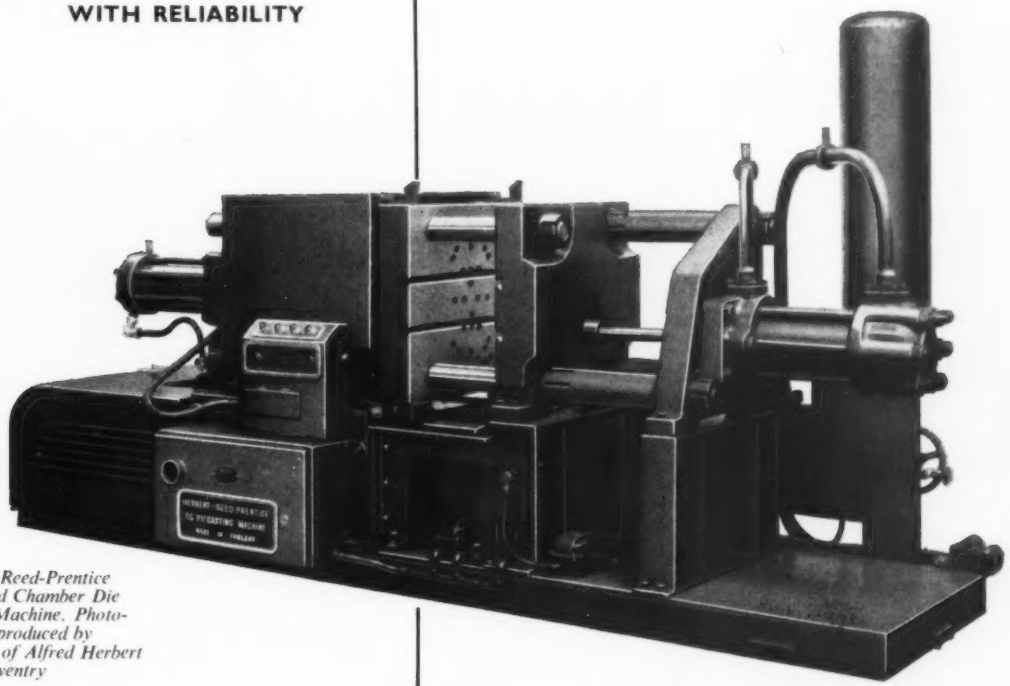

Herbert-Reed-Prentice 1½G Cold Chamber Die casting Machine. Photograph reproduced by courtesy of Alfred Herbert Ltd., Coventry

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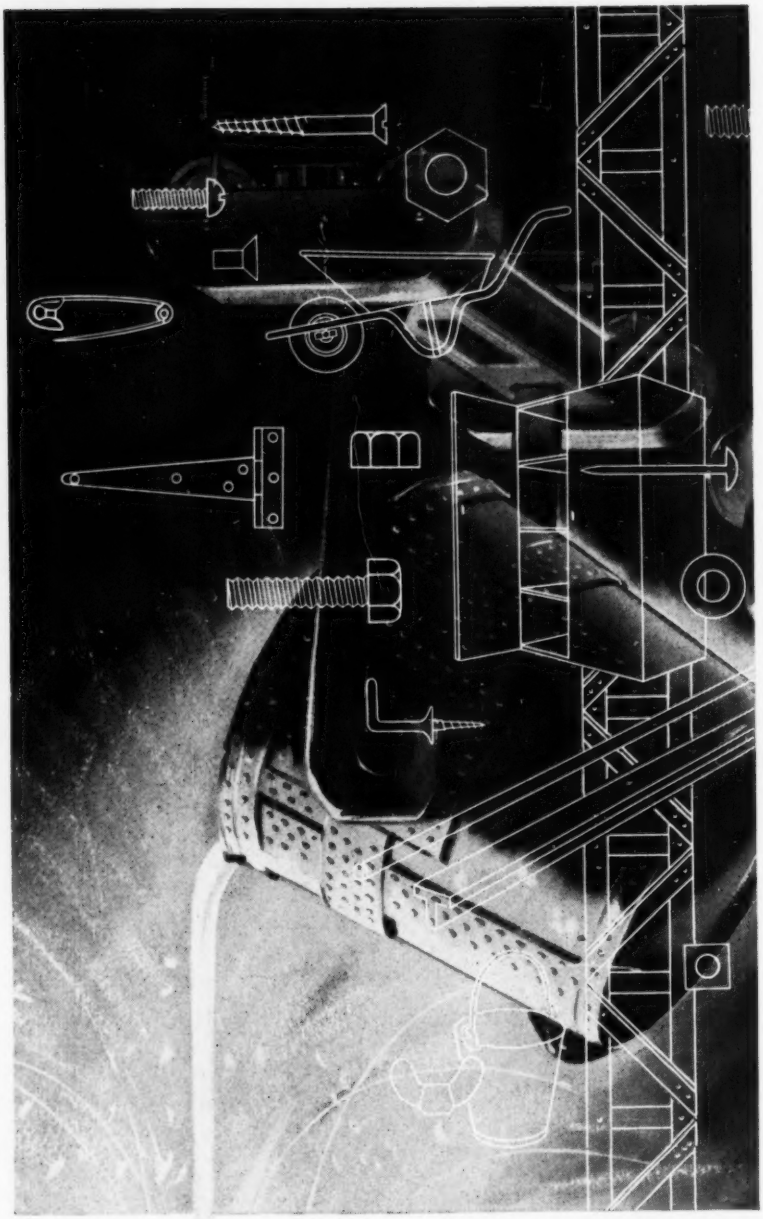

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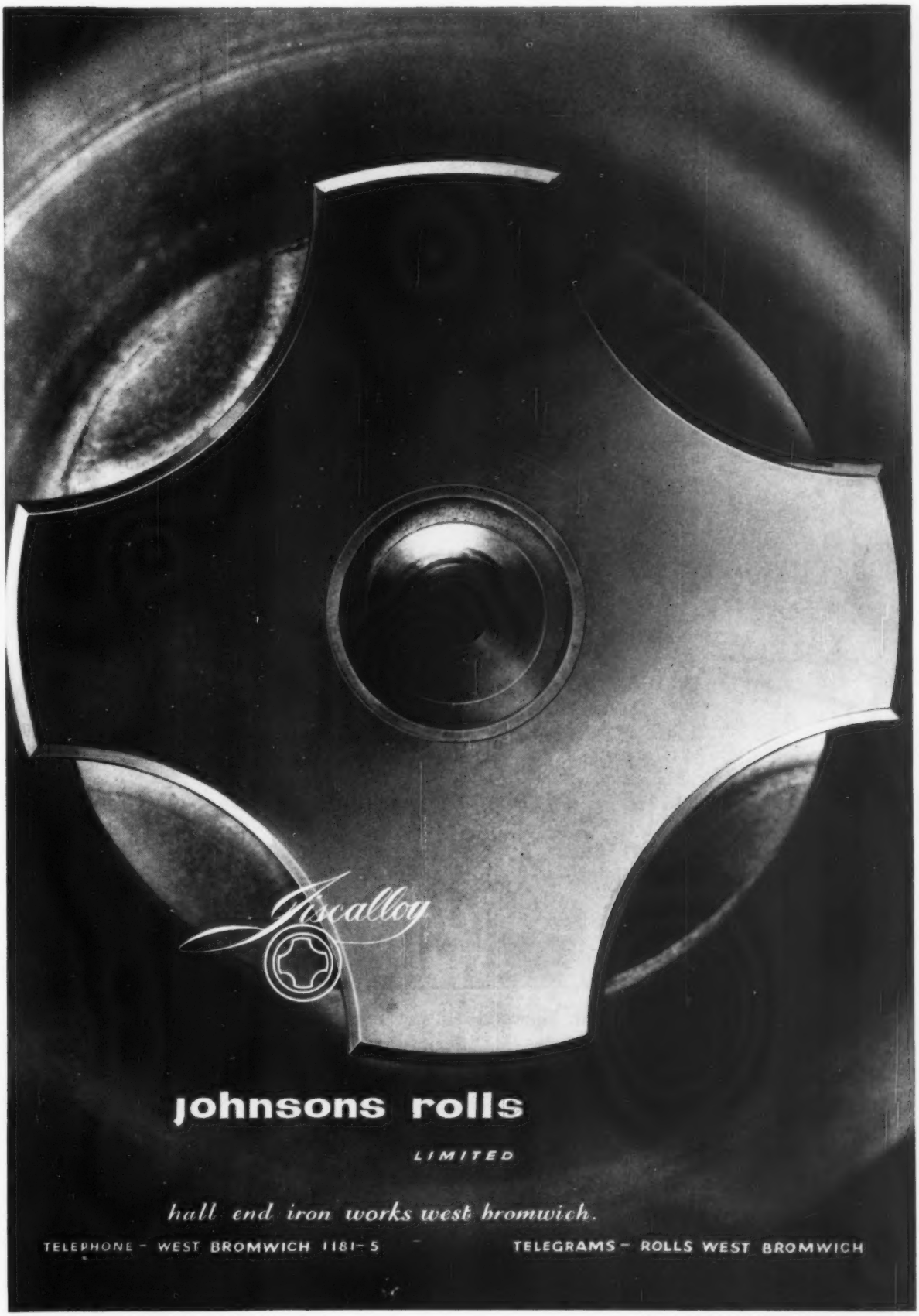
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This complete BIGWOOD slitting line operates in the Castle Works of Guest Keen & Nettlefolds (South Wales) Ltd

BIGWOOD

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The line can deal with strip up to a maximum width of 20 inches in thicknesses from 0.04 to 0.064 ins. at a pass speed of 250/450 ft. per minute.

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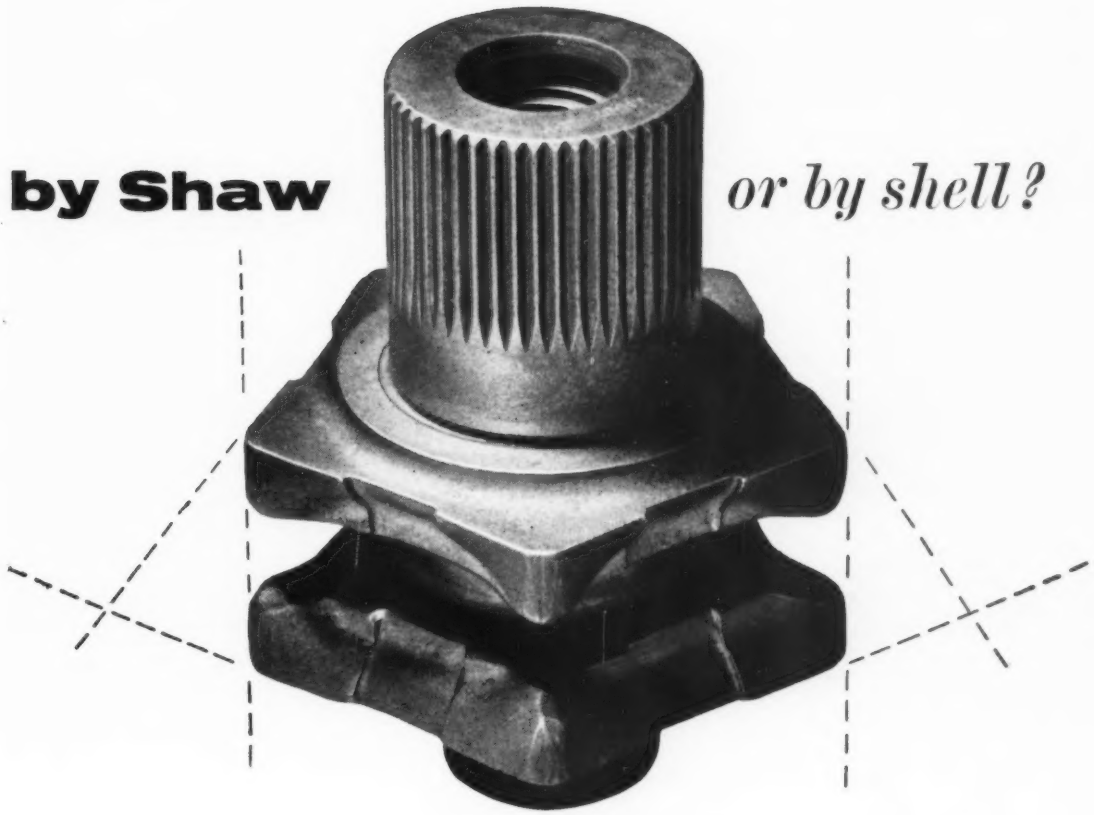
B.142

by Shaw *or by shell?*



As licensees of the Shaw Process, and producers of castings by the Shell Moulded Process as well, The PHOSPHOR BRONZE Co. Ltd. can give you an unbiased opinion on the best method to use in any instance. While castings produced by the shell moulded process are more accurately produced than by older methods, the limits cannot be held to the same high degree of accuracy as when castings are produced by the Shaw Process.

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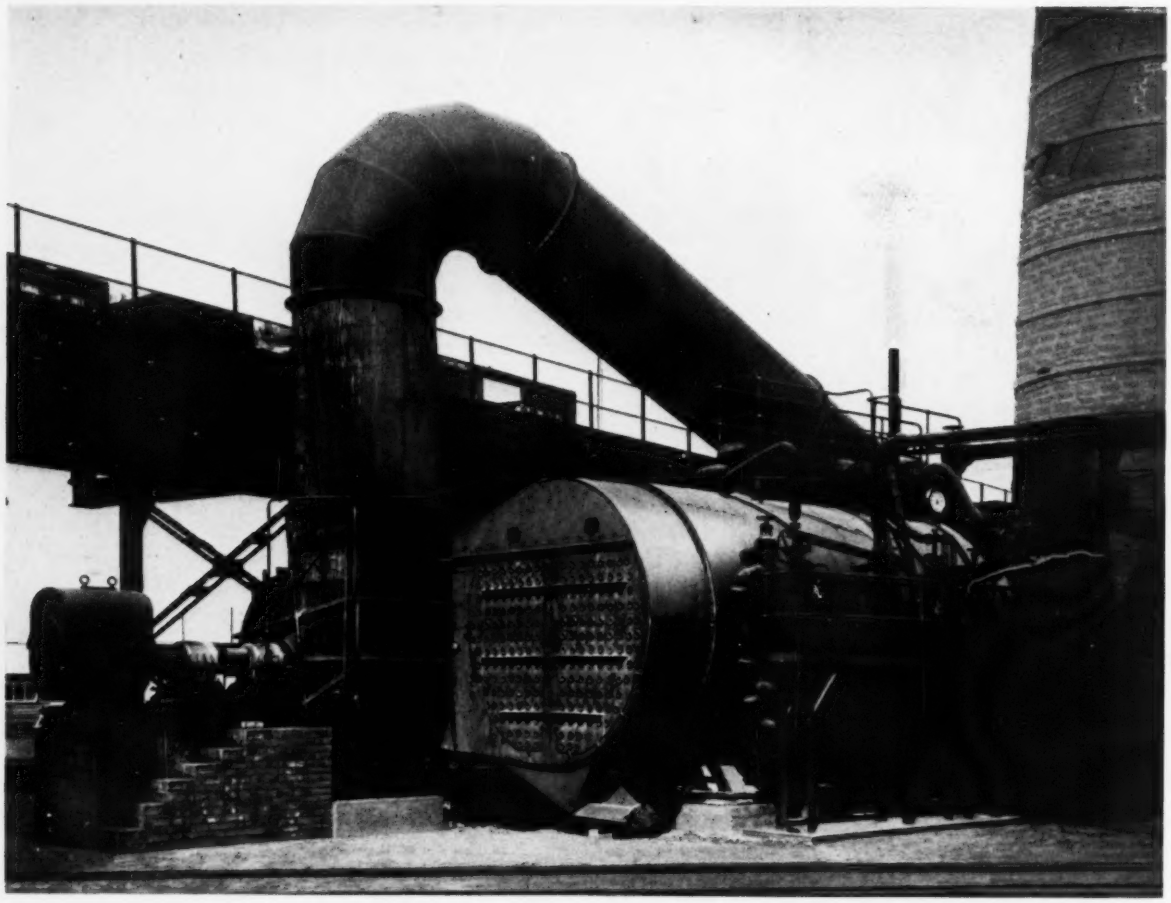

Photograph by courtesy of Messrs. John Lysaght's Scunthorpe Works Ltd.

One of a battery of SPENCER-BONECOURT Waste Heat Boilers operating on Open Hearth Steel Melting Furnaces at John Lysaght's Scunthorpe Works Ltd., Scunthorpe.

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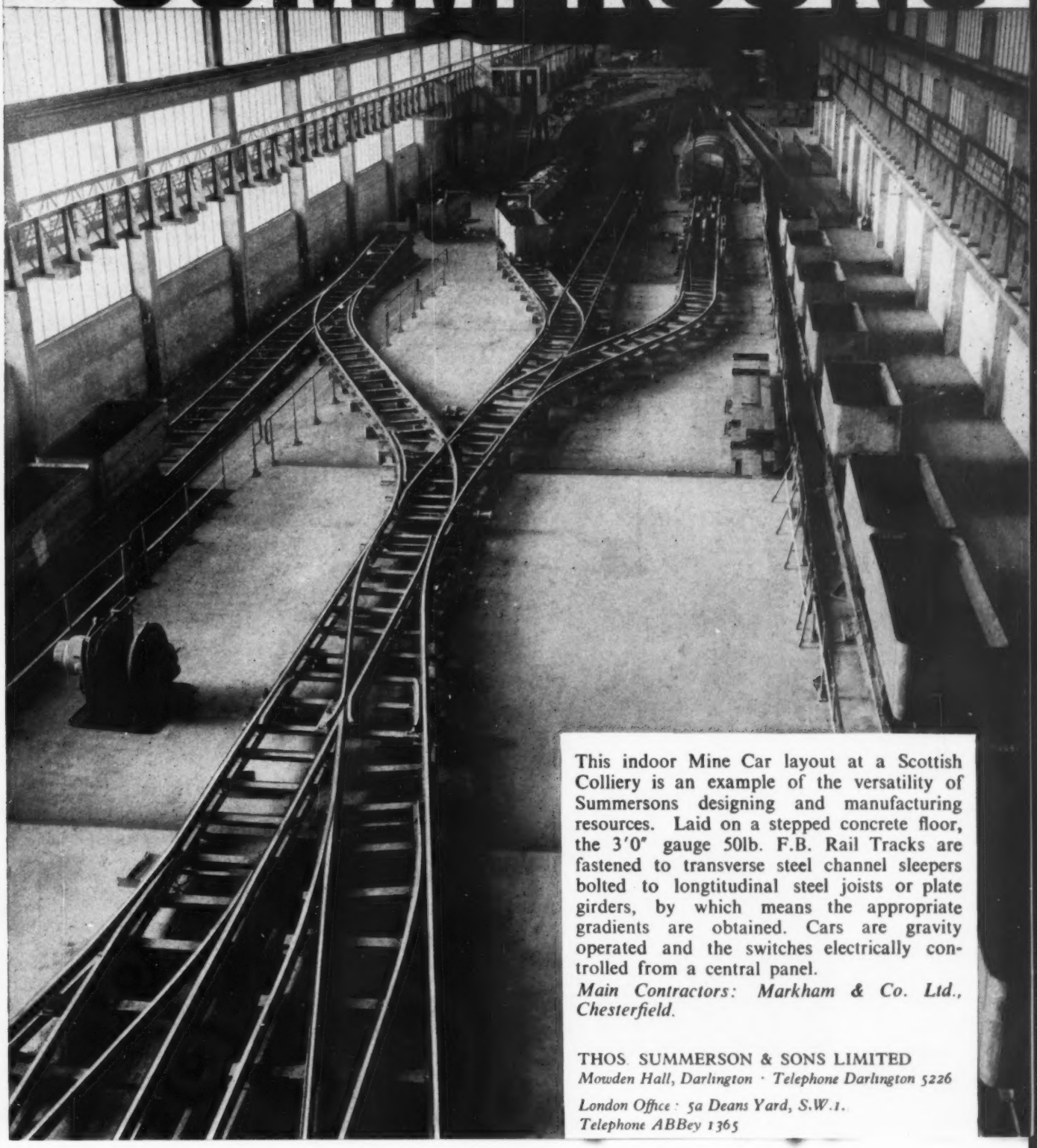

This indoor Mine Car layout at a Scottish Colliery is an example of the versatility of Summersons designing and manufacturing resources. Laid on a stepped concrete floor, the 3'0" gauge 50lb. F.B. Rail Tracks are fastened to transverse steel channel sleepers bolted to longtitudinal steel joists or plate girders, by which means the appropriate gradients are obtained. Cars are gravity operated and the switches electrically controlled from a central panel.

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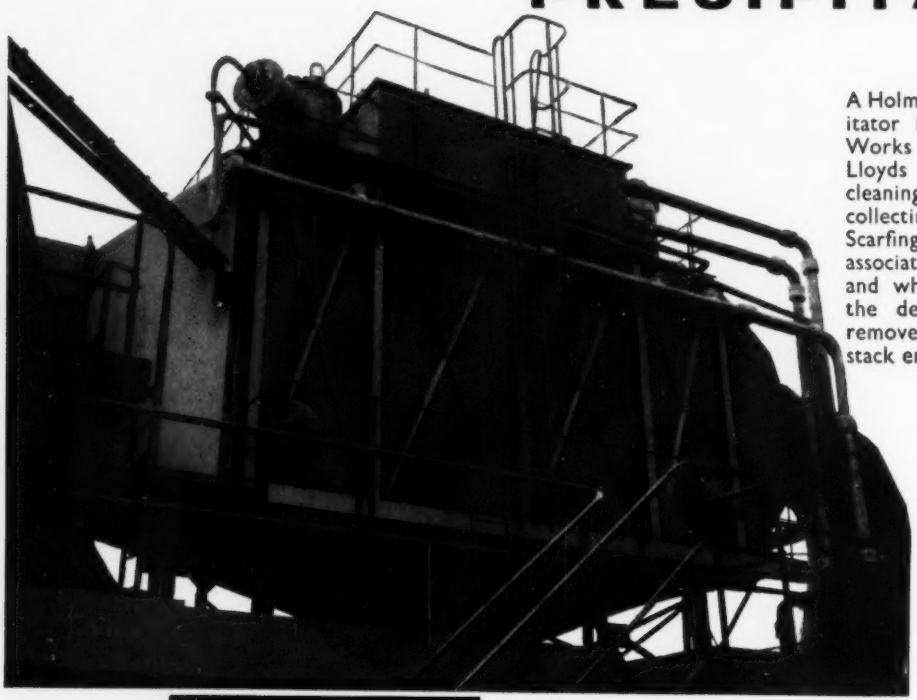

A Holmes-Elex Electrical Precipitator installed at the Corby Works of Messrs. Stewarts & Lloyds Ltd., for the purpose of cleaning exhaust air from the collecting hood enclosing the Scarfing Plant. The fine particles associated with iron oxide fume, and which are responsible for the dense brown colour are removed resulting in no visible stack emission.

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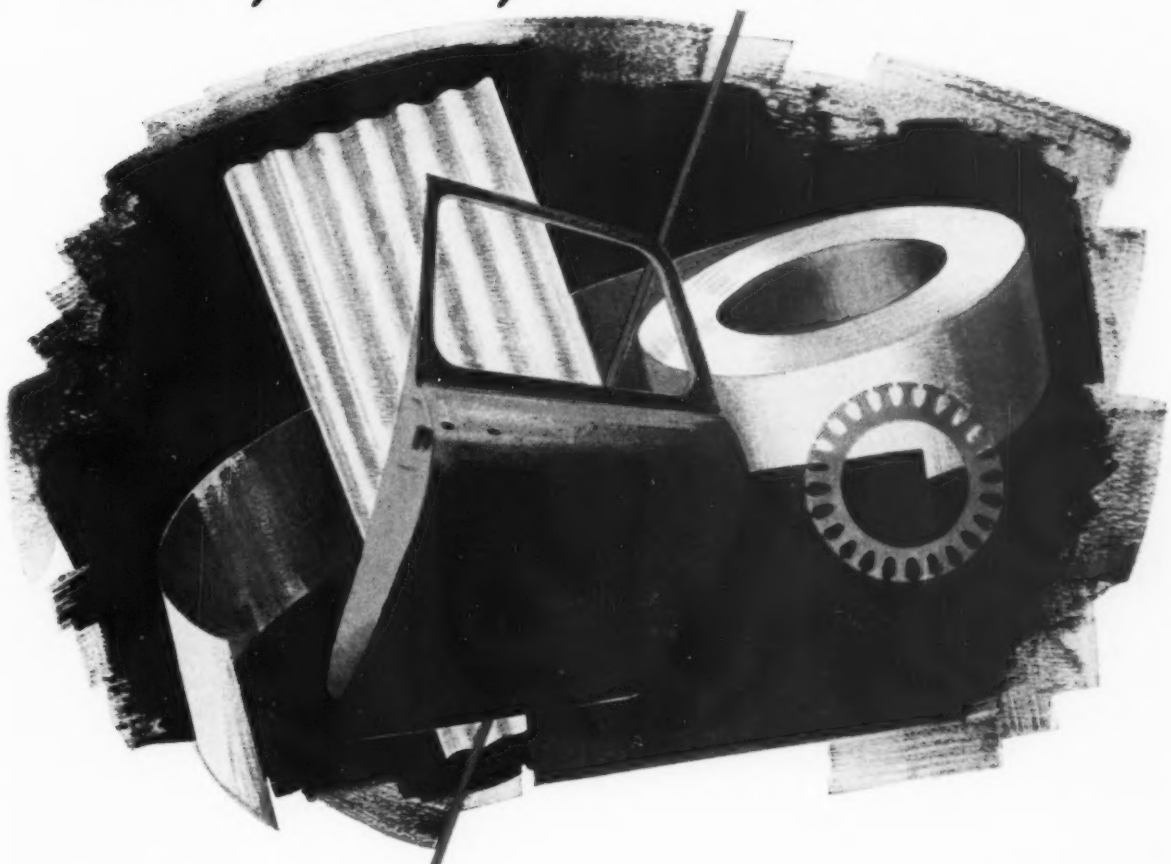

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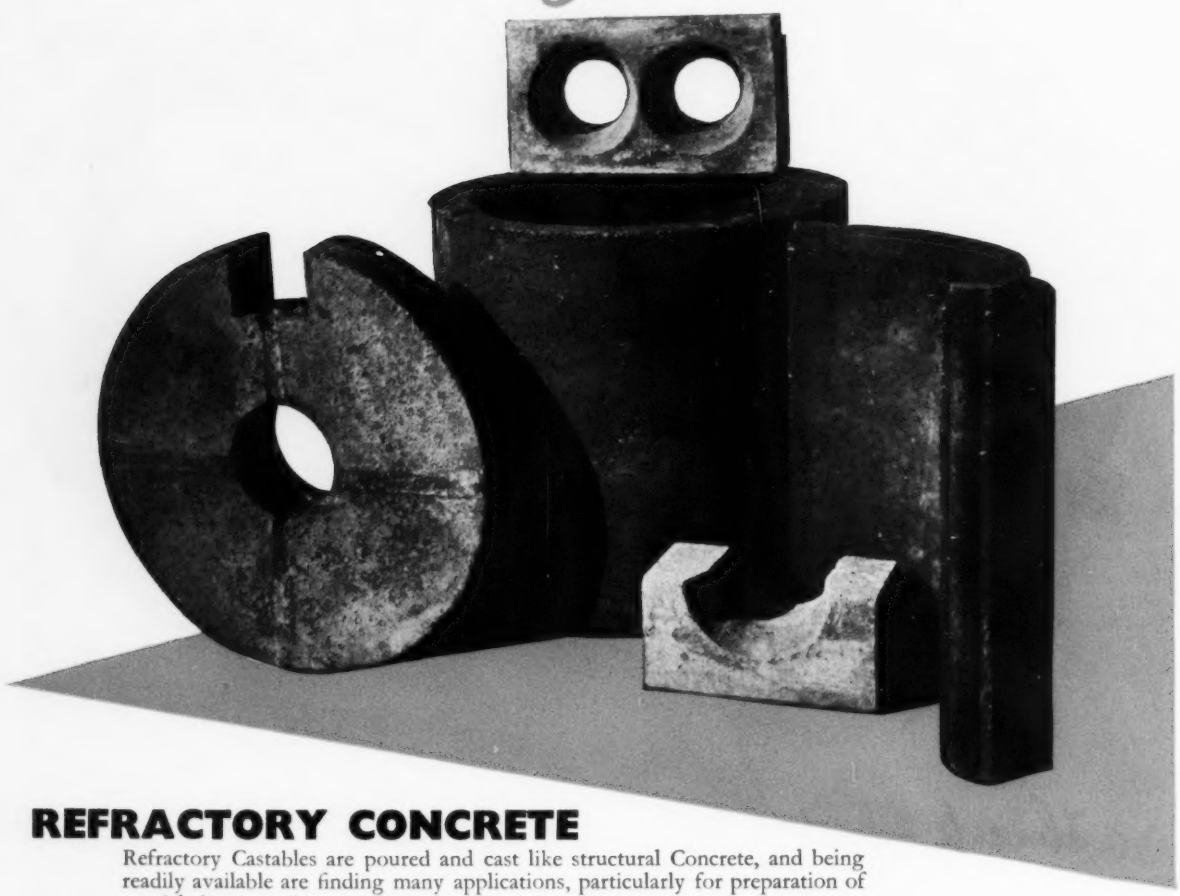

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Stein Super Refractory Concrete No. 16	Dry	+1750°C	Hydraulic	1200°C	1600°C	160
No. 17	Dry	+1750°C	Hydraulic	1200°C	1700°C	160
No. 18	Dry	+1750°C	Hydraulic	1200°C	1800°C	160
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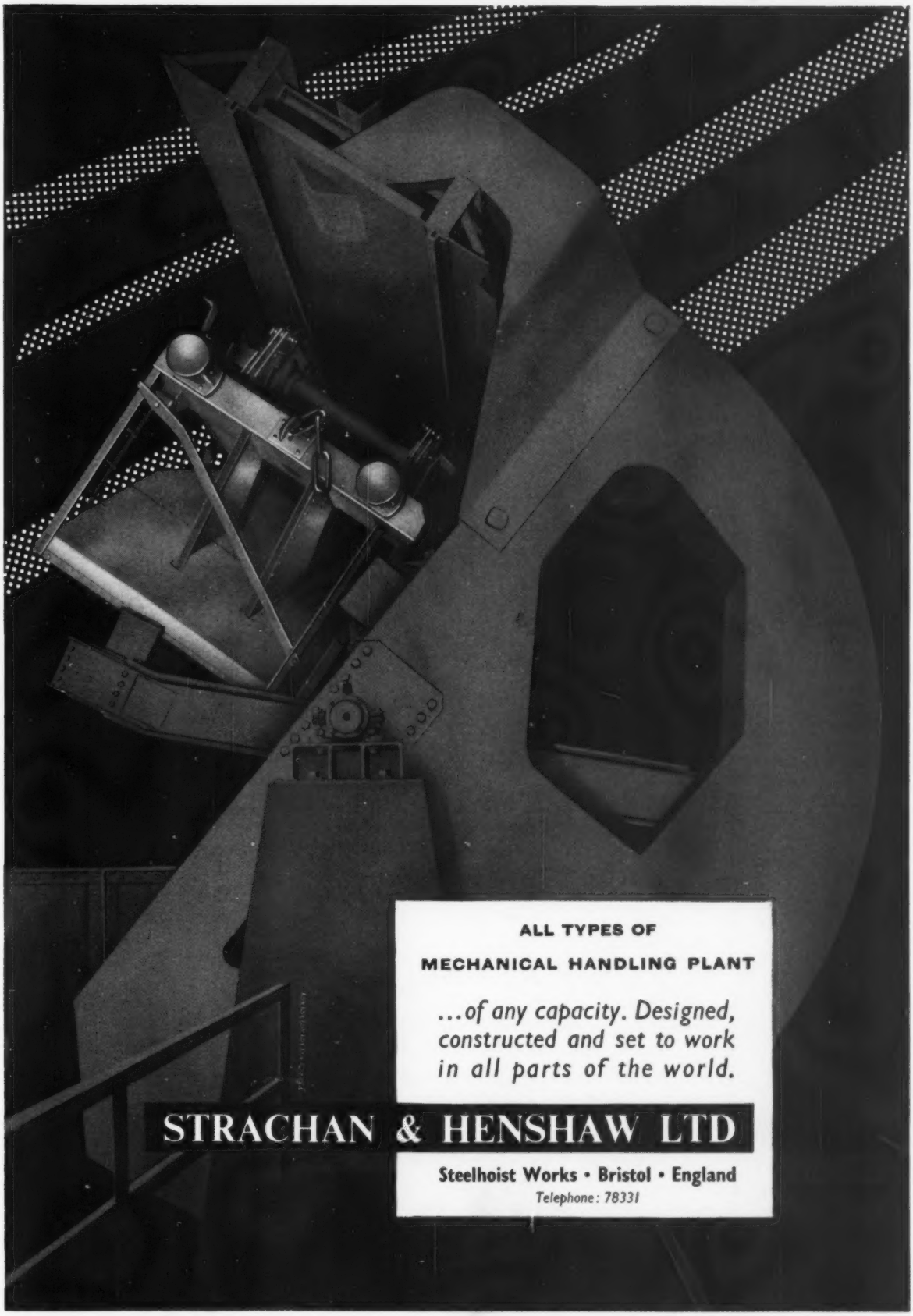
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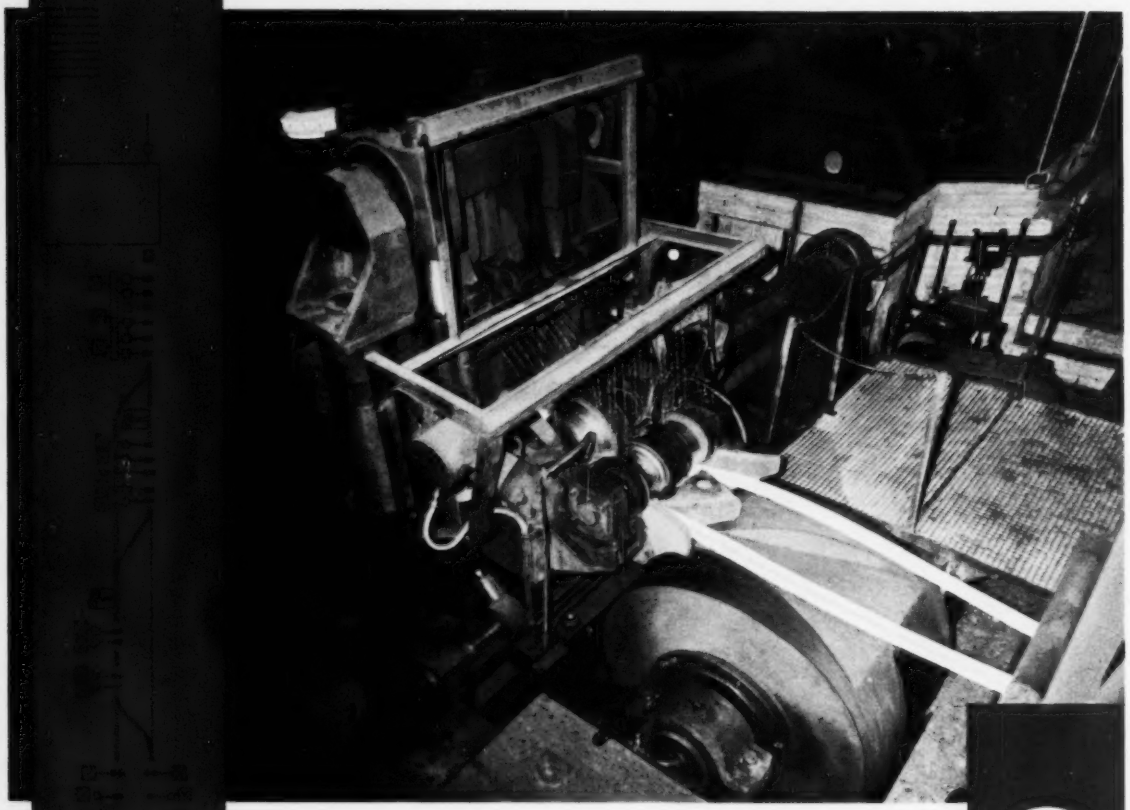

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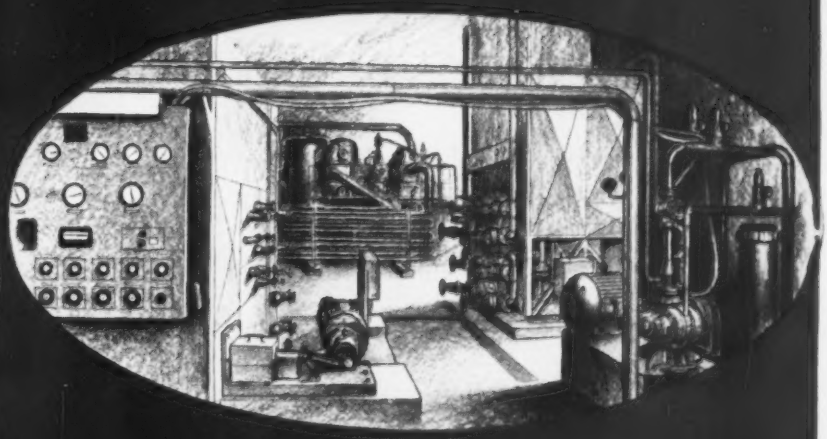

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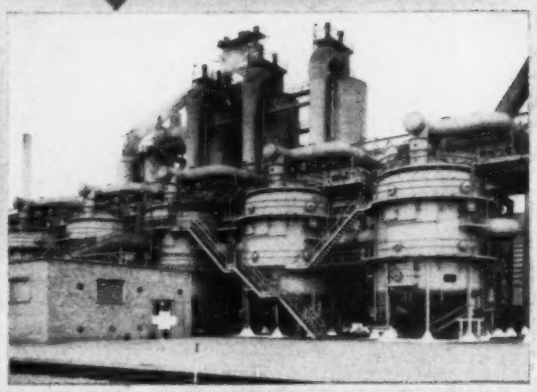
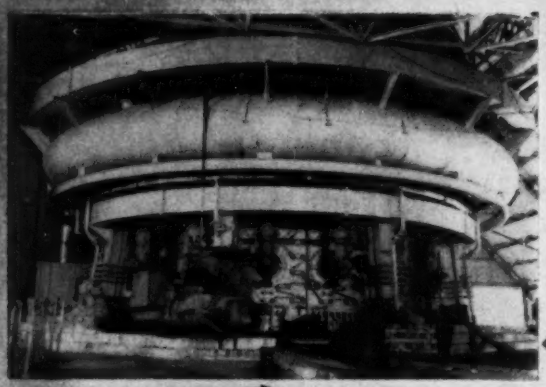
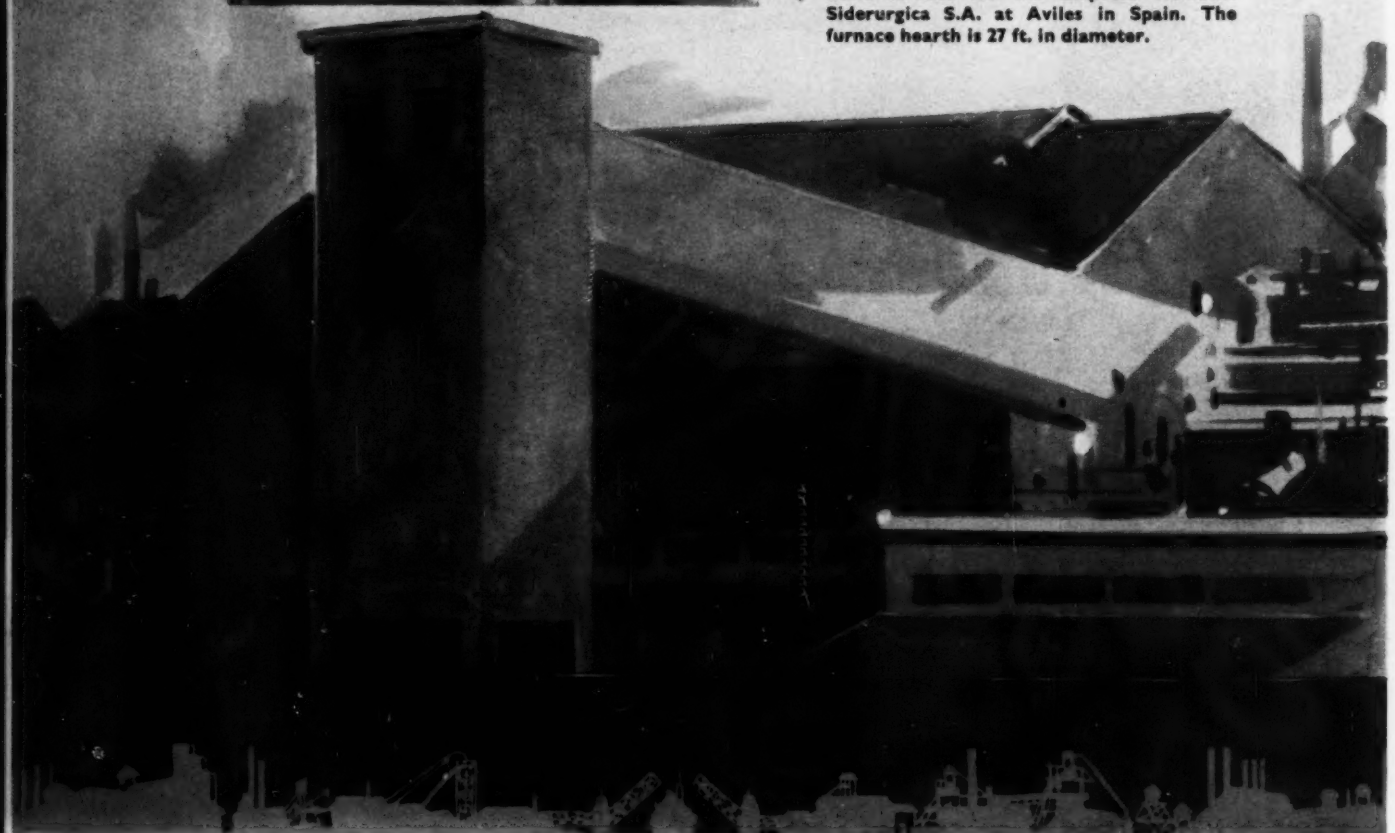




Blast furnace for the Empresa Nacional Siderurgica S.A. at Aviles in Spain. The furnace hearth is 27 ft. in diameter.



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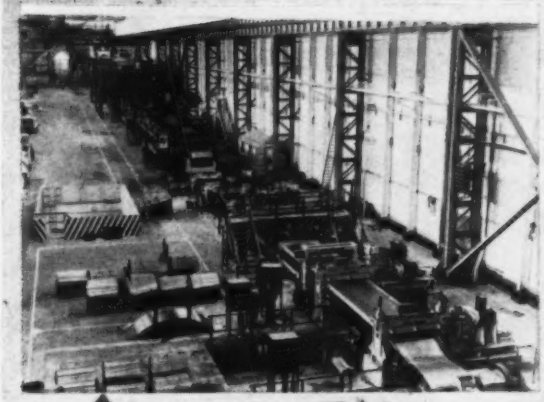

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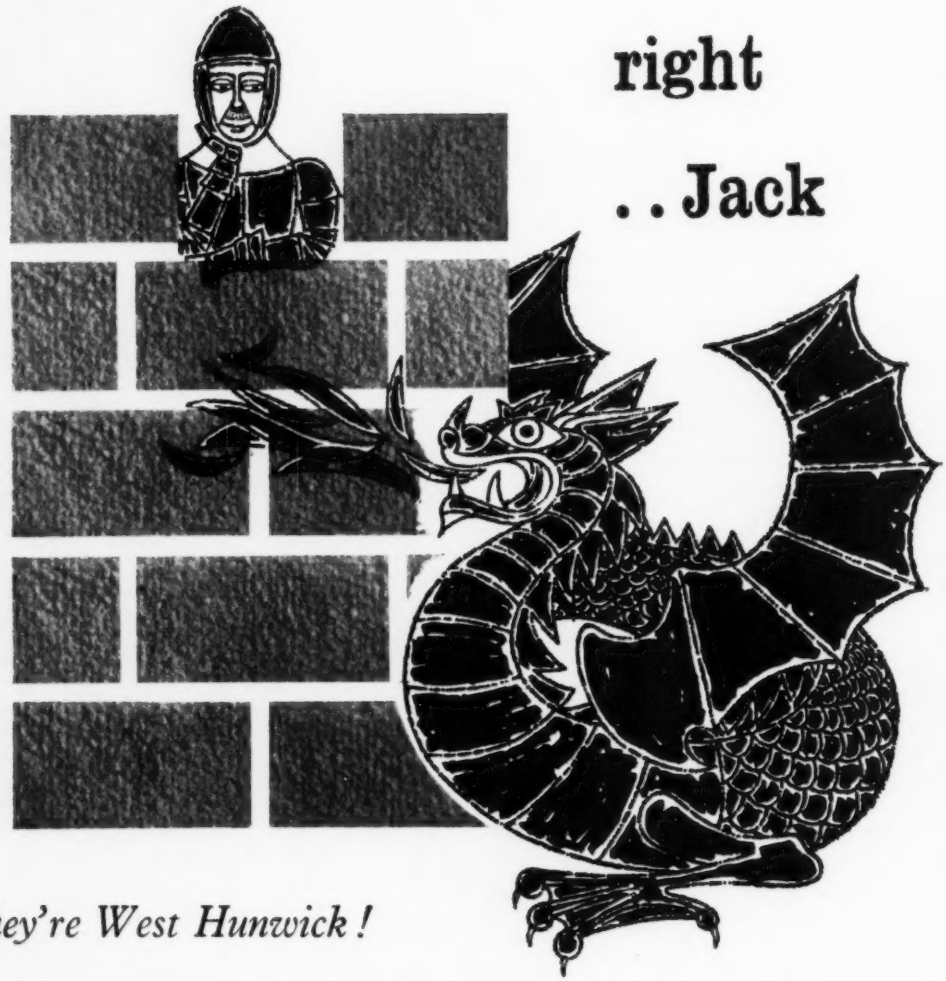
I'm all right ..Jack



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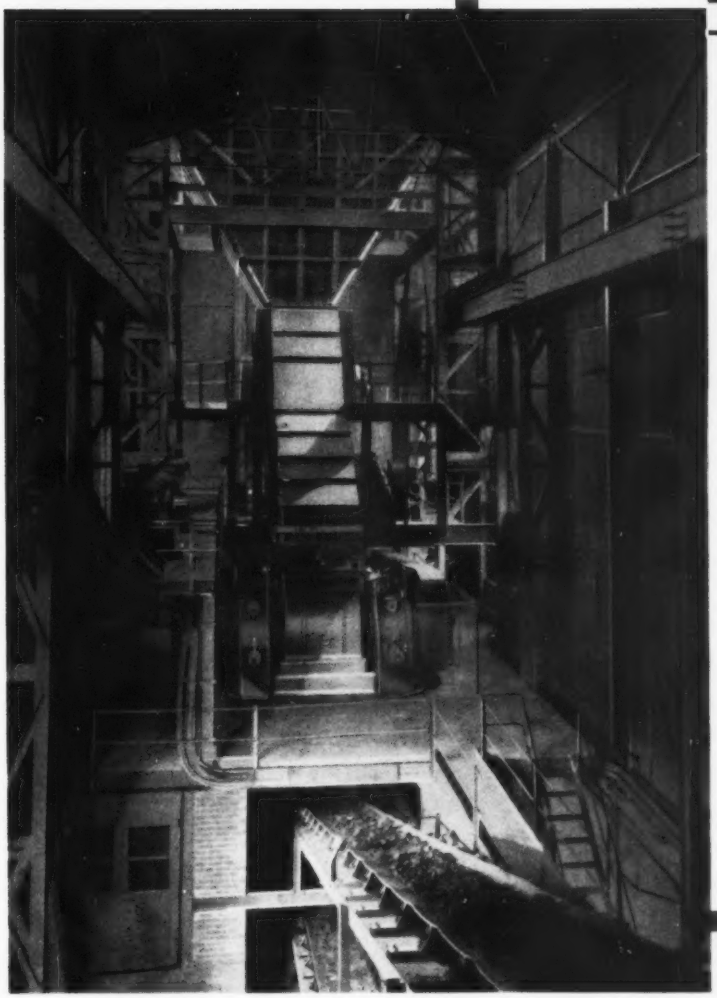

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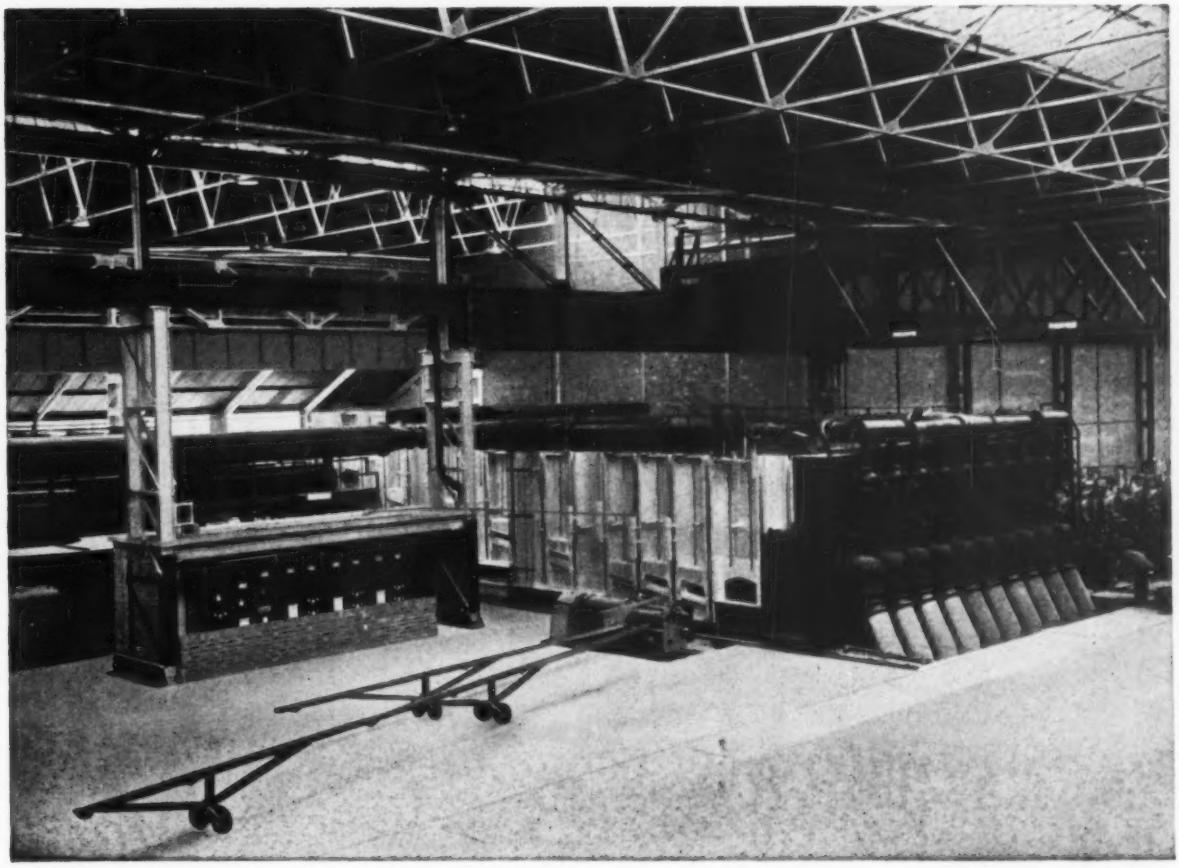

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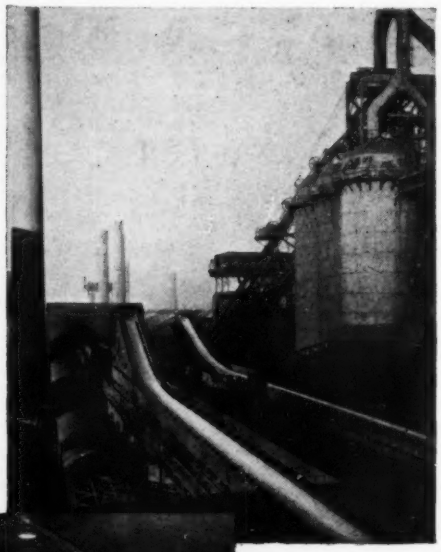

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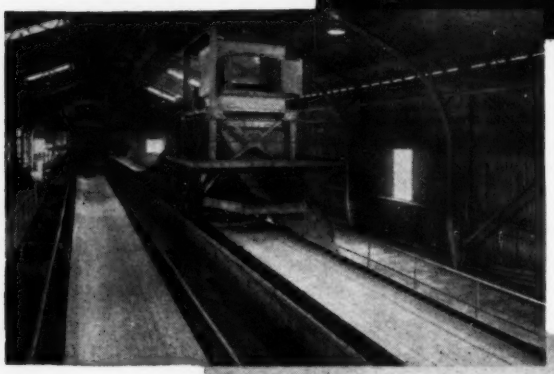
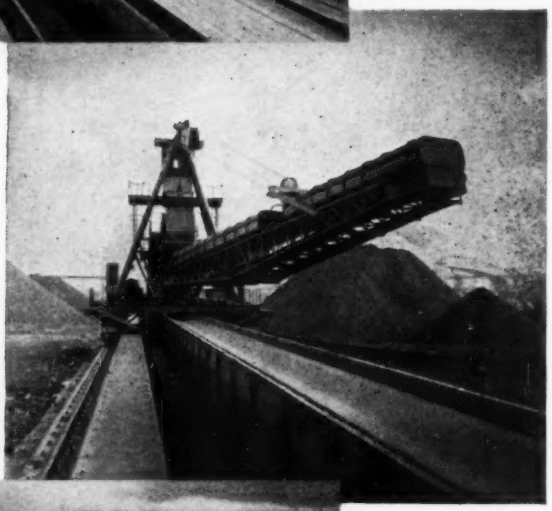


Trippers feeding Ore to Dockside Bunkers.



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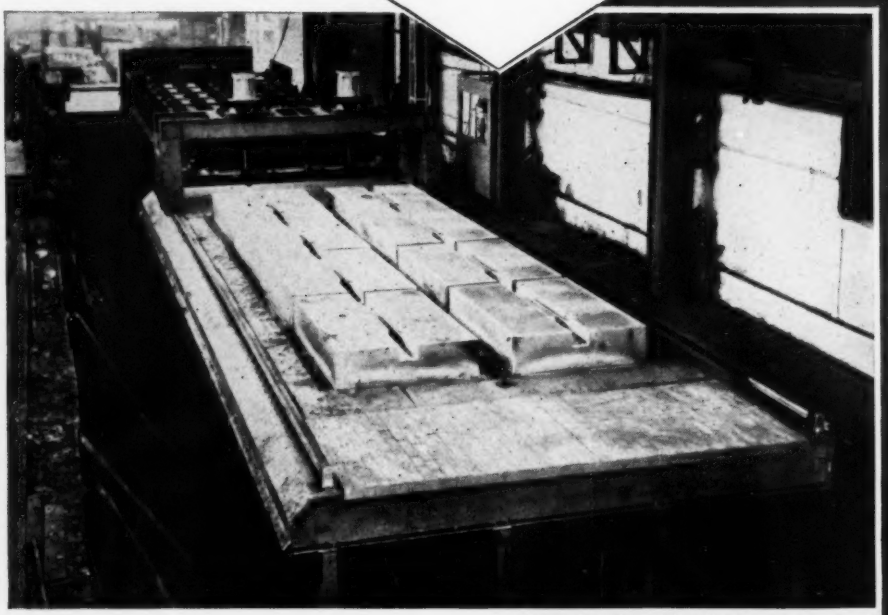
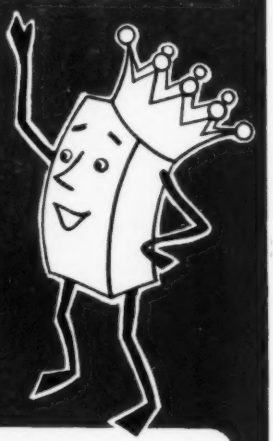


Photograph reproduced by courtesy of Appleby — Frodingham Steel Company (Branch of The United Steel Companies, Ltd.) and Stein and Atkinson, Ltd.



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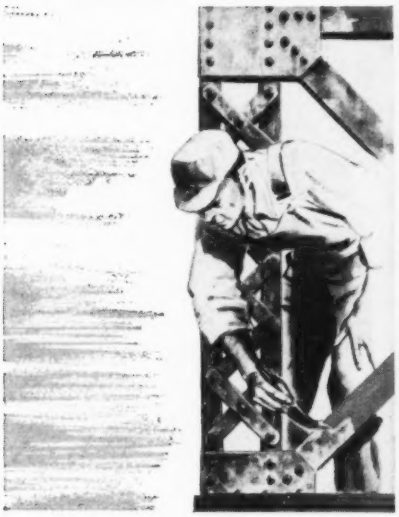

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- ☐ Please have your Representative call to demonstrate Rust-Oleum.

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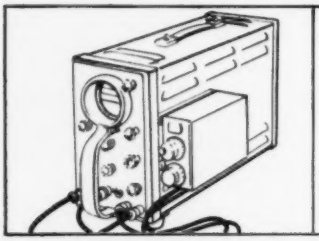
Choose your equipment from this new comprehensive range

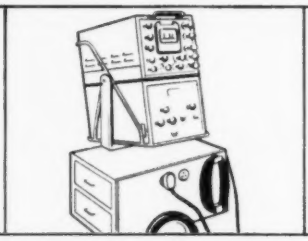
Kelvin Hughes ultrasonic flaw detection instruments and probes are available for all manual, semi-automatic and automatic testing techniques. High performance and versatility are obtainable with this comprehensive range of competitively priced equipment designed for laboratory work as well as routine inspection schedules. Auxiliary units include Automatic Flaw Alarms, Depth Gauges, Probe Holders and Manipulators.

ULTRASONIC FLAW DETECTOR Mk. 6

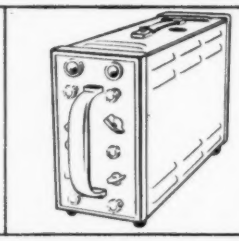
This powerful, versatile instrument has an exceptionally wide operating frequency range, and a variety of controls enabling particular methods of tests to be devised. The trolley mounted display and control units are readily adjustable to the most convenient position. A brilliant trace is given over the full width of a 6 in. C.R.T. Operates with single, twin or separate probes for contact, gap or immersion scanning. Automatic flaw alarm circuits, depth gauge and photographic recording facilities can be added.



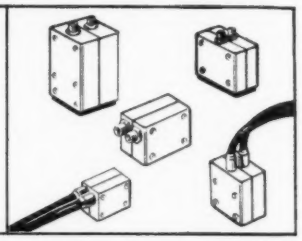
Flaw Detectors Marks 5F, 5AF. *Portable instruments with similar basic specifications. The 5F is a multi-frequency instrument for use with twin or separate type probes, whilst the 5AF, a single frequency instrument, either 1½, 2½ or 5 Mc/s. Also operates with single type probes. (Shown with Depth Gauge attachment).*



Flaw Detector Mark 6. *Trolley mounted transportable equipment, very powerful and versatile for use with all types of probes and having an exceptionally wide frequency range and high sensitivity.*



Two Channel Flaw Alarm. *For use with Mark 5 instruments, a portable auxiliary unit providing semi-automatic monitoring facilities which identify and give warning of significant defects.*



Probes. *Extensive development by Kelvin Hughes has resulted in a range of standard probes having either single, twin or separate transducers, and a very high performance throughout.*

AUXILIARY EQUIPMENT

Includes a Depth Gauge for use with all instruments and which will measure thicknesses between 0.1 to 4 ins. with an accuracy of 0.75%, Probe Holders with irrigation facilities for large surface scanning, and remote Probe Manipulators.

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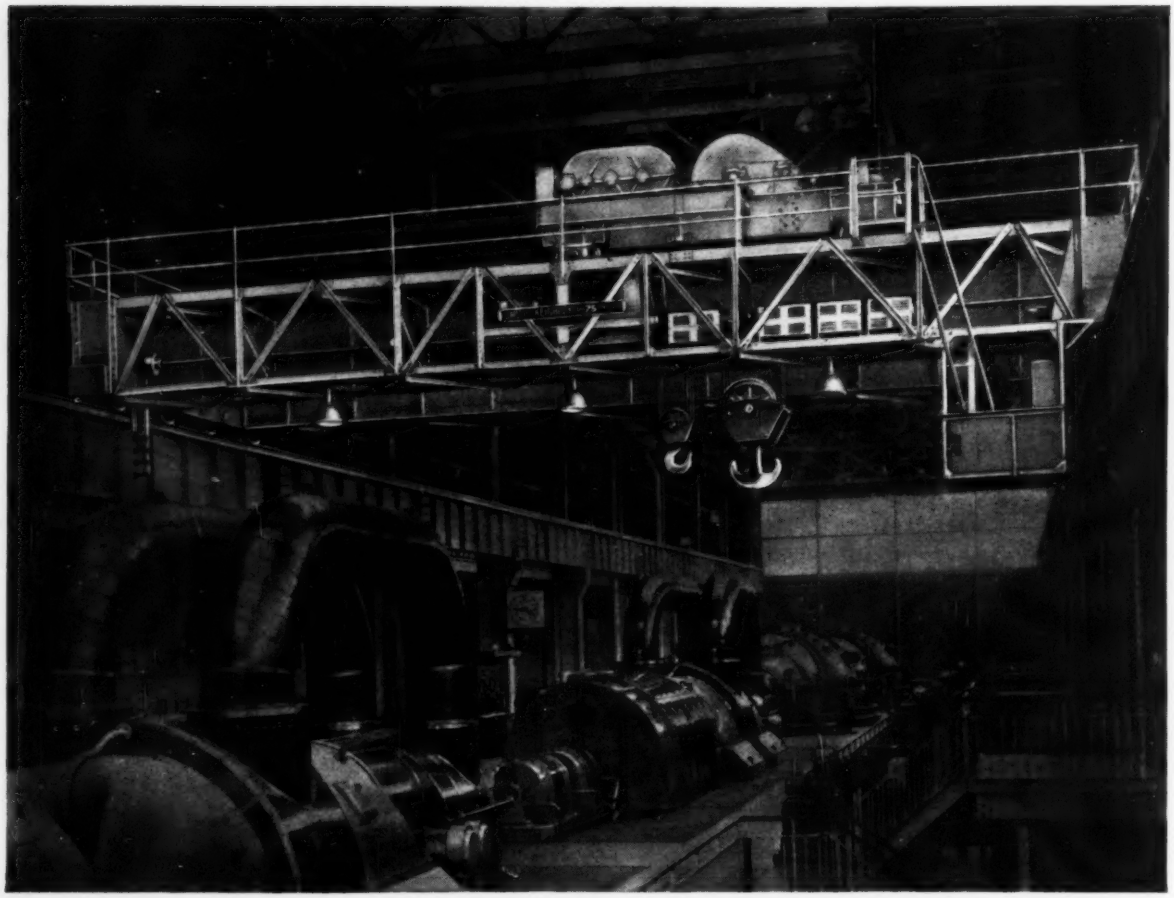

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Under a comprehensive contract, 'ENGLISH ELECTRIC' have installed the major portion of the electrical equipment for a slabbing mill in the largest steelworks in Australia. The twin-drive is rated at 12,000 h.p. (R.M.S.) at 50/80 r.p.m., with a cut-out torque of 514 metre tons, and is capable of a base speed to base speed reversal *in under one second*. The mill follows the modern trend in having individual driven rolls on the front and back mill tables totalling 21 in all. A total of 60,000 h.p. of electric motors has been supplied which includes the main drive, a 9,000 h.p. A.C. motor for the Ilgner Set, four Ward Leonard Auxiliary M.G. Sets and associated control. 'ENGLISH ELECTRIC' have more than 50 years of world wide experience in steelworks electrification which includes installation of the first twin-drive in Europe over 25 years ago.

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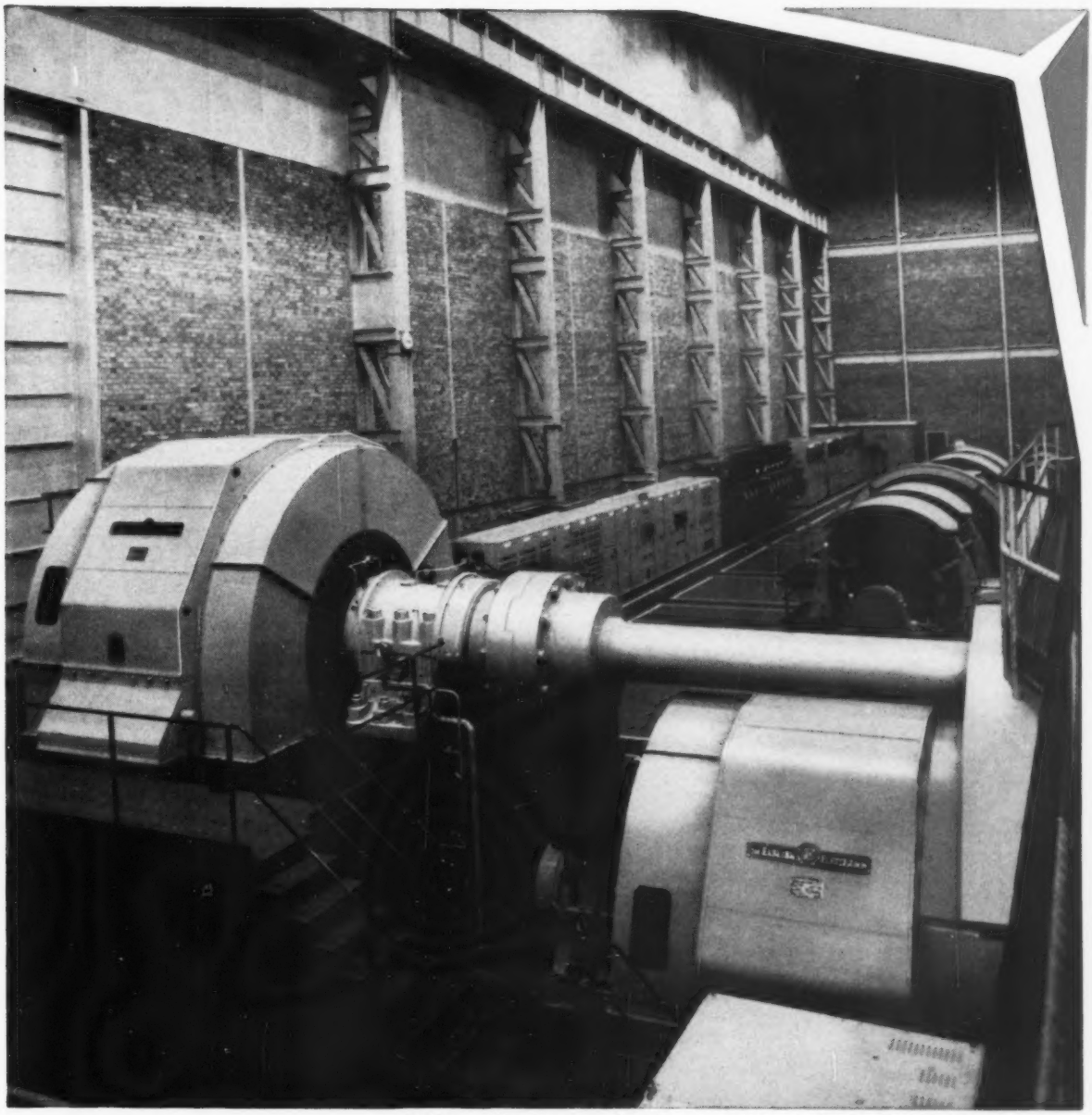

Slabbing mill motor room of an Australian steelworks housing the twin-drive motors

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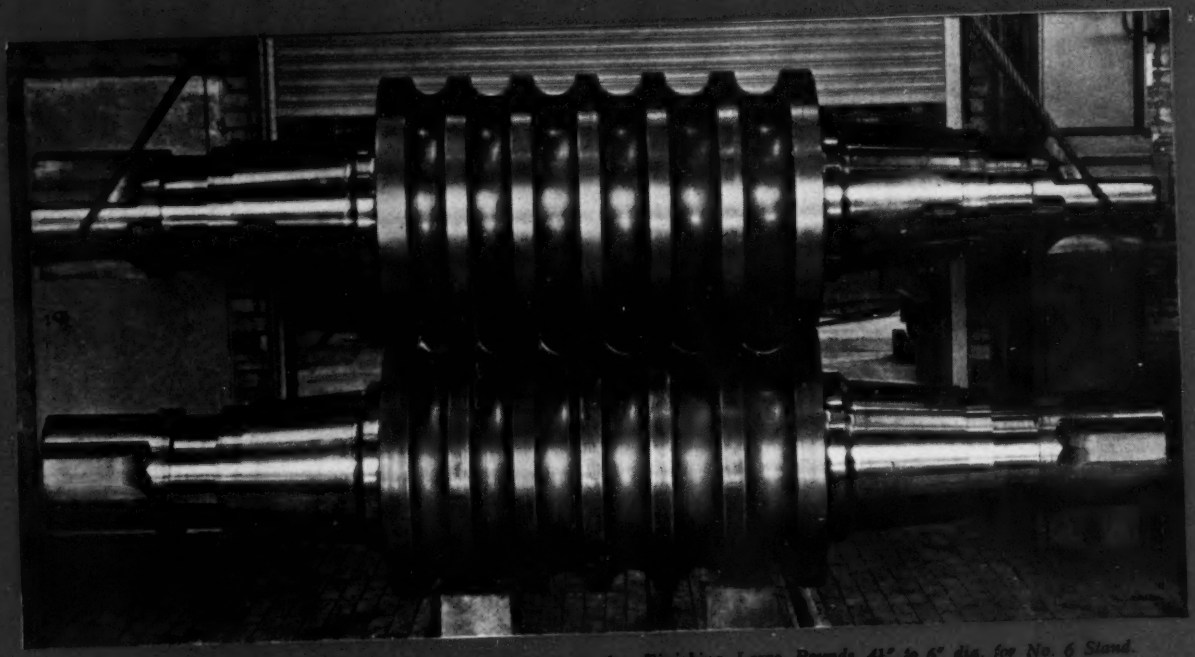

A pair of our Special "SSXH" Moorralloy Quality Rolls for Finishing Large Rounds 4½" to 6" dia. for No. 6 Stand. Roll size 32" dia. x 54" long barrel.

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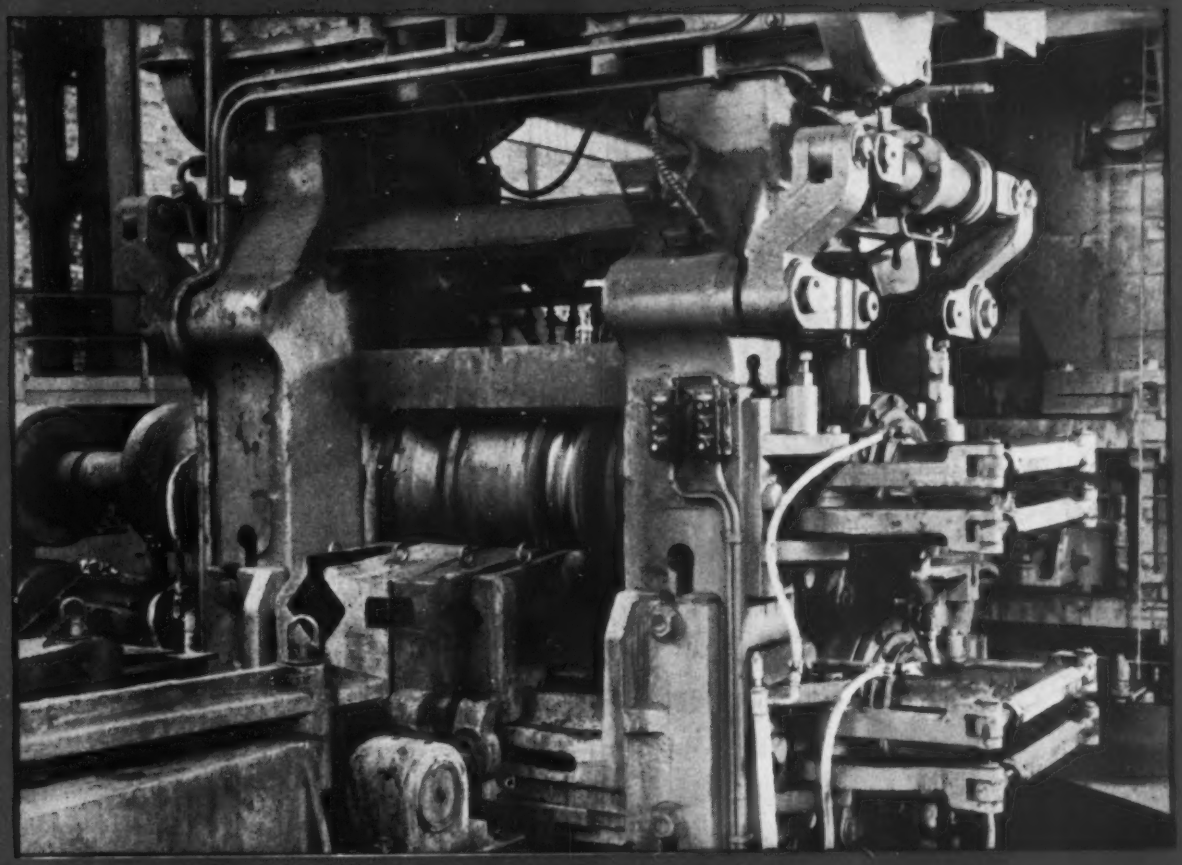

No. 6 Stand Finishing for 4", 4½" and 5" billets, also Roughing for 2" to 3½" Billets and Slabs at Consett Iron Co., Ltd., showing our Special S.S.X. Mooralloy Quality Rolls. Roll size 32" dia. x 54" long barrel.

Photographed by the courtesy of The Consett Iron Co. Ltd.

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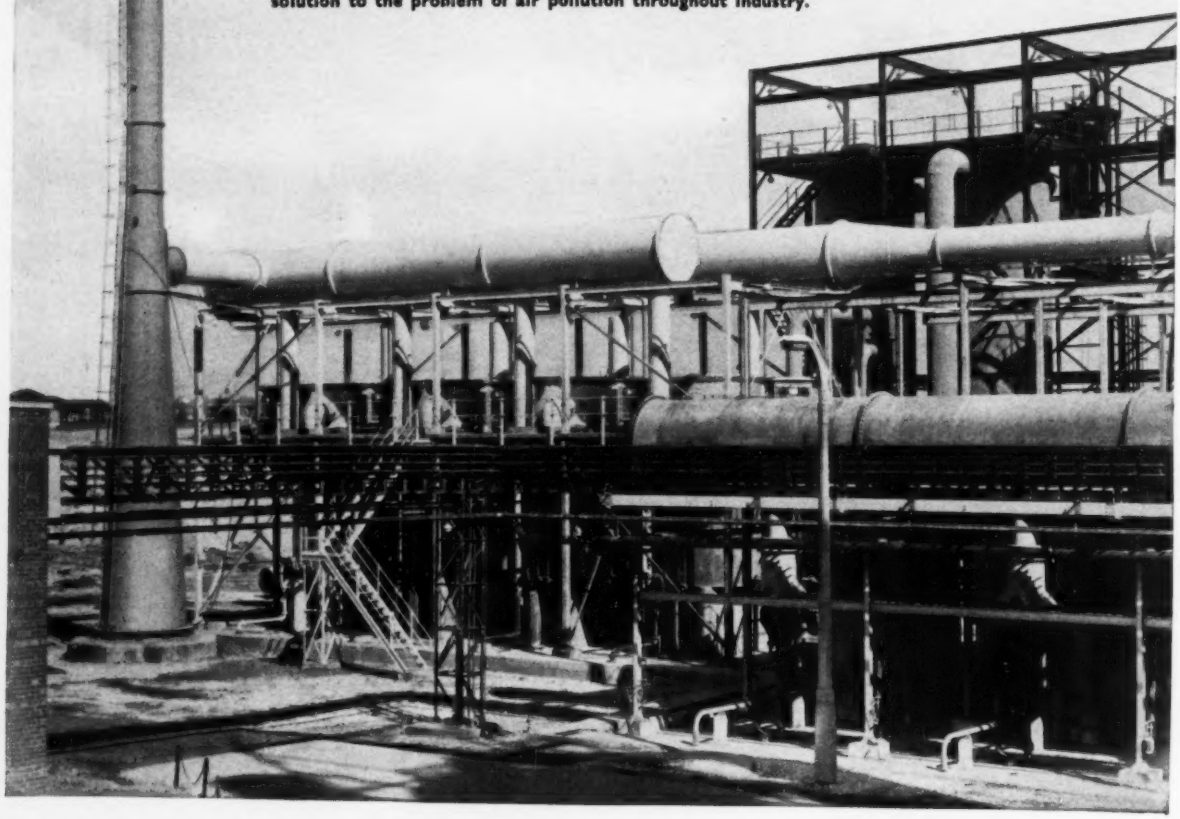

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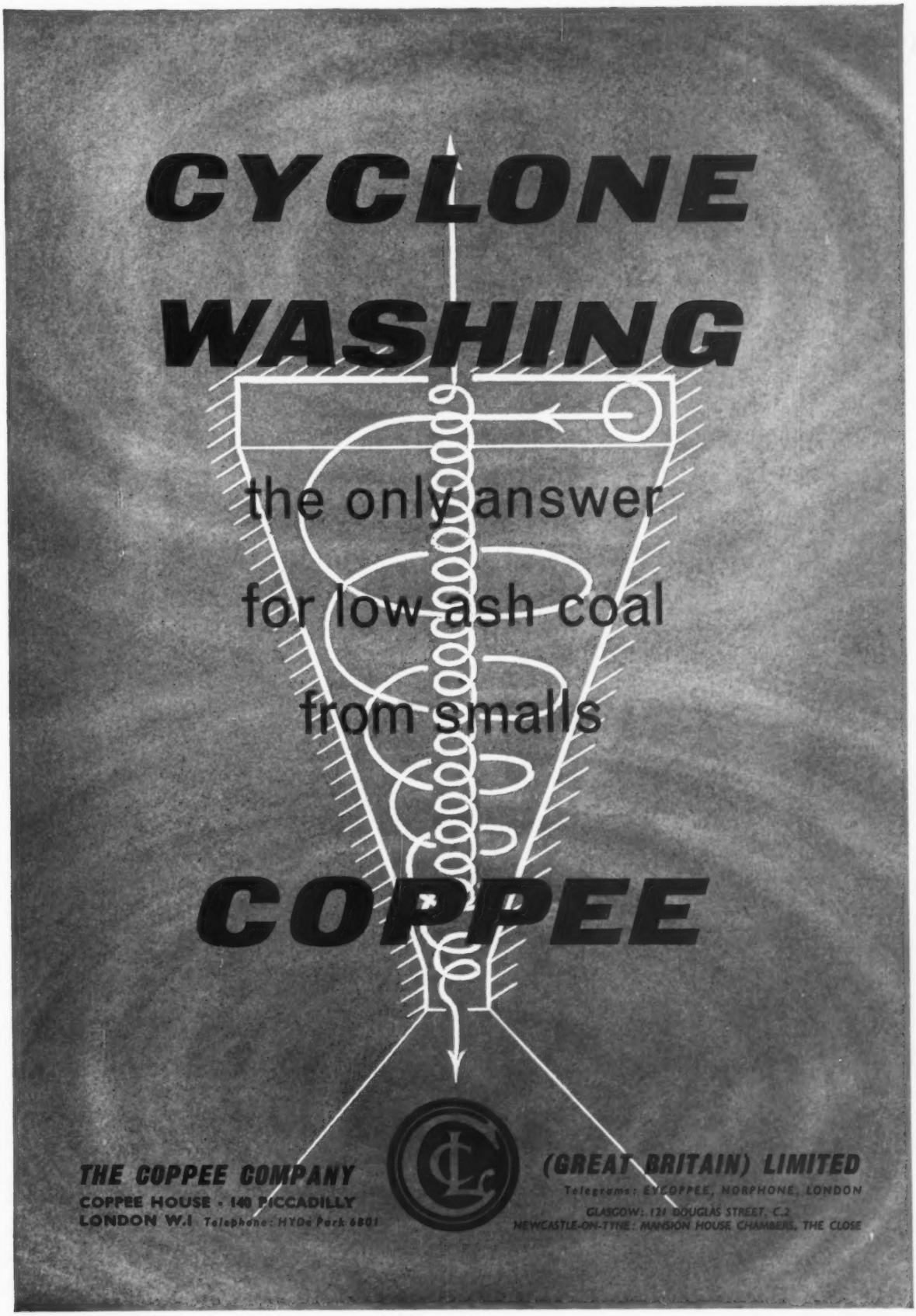
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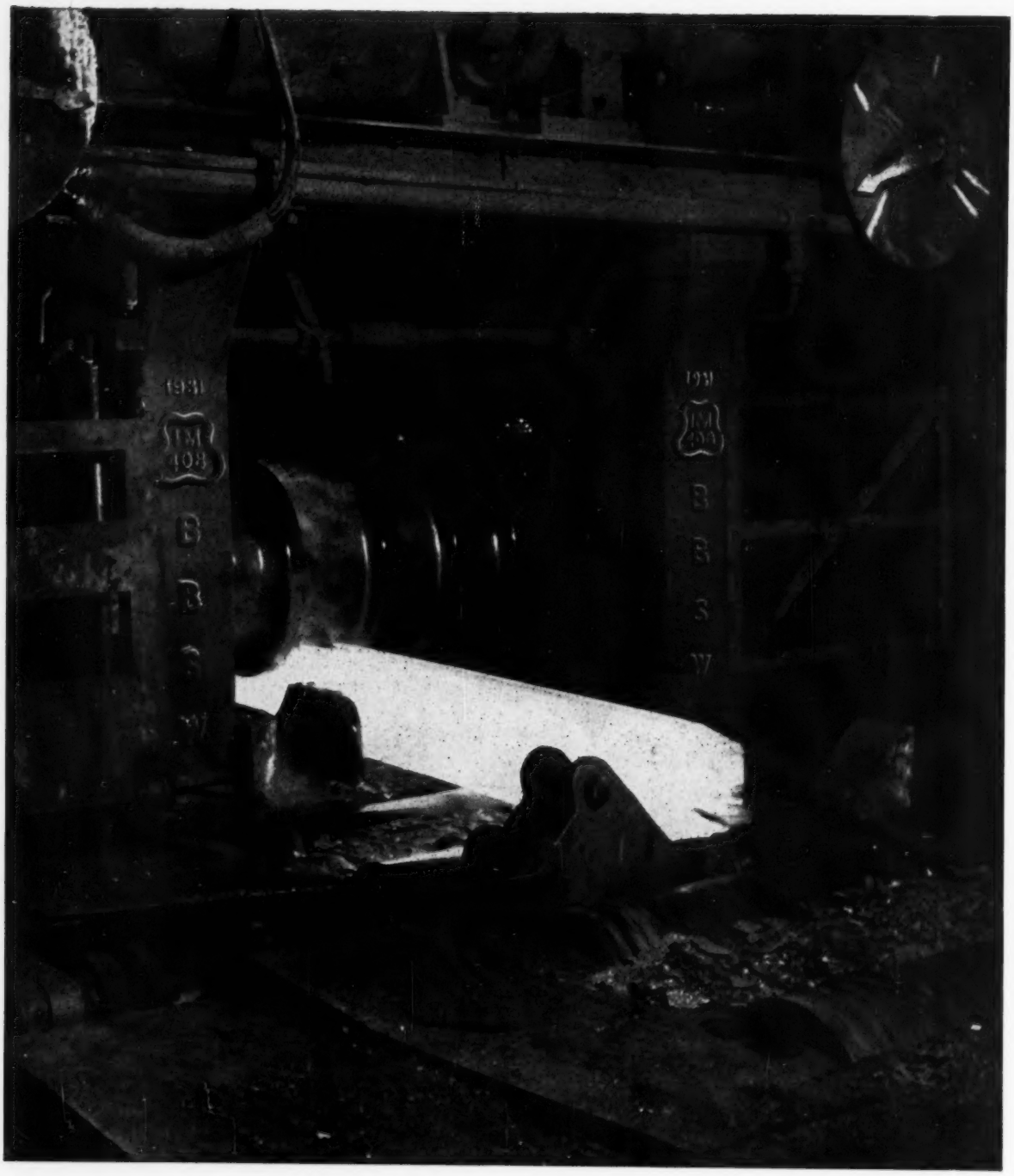
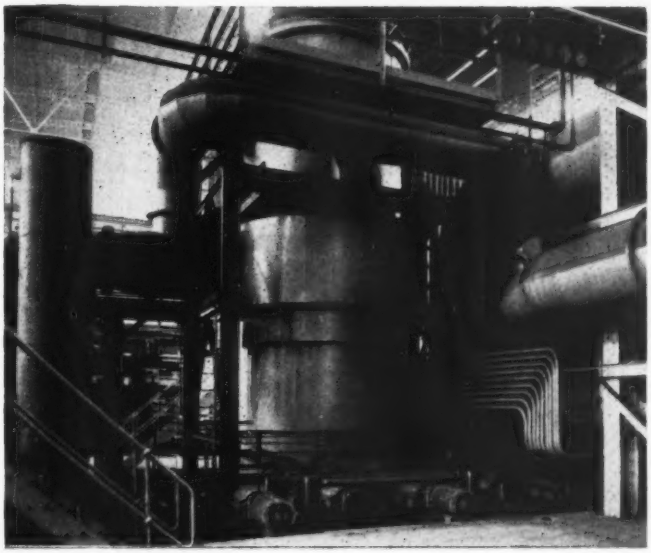


BROWN BAYLEY STEELS LIMITED SHEFFIELD

M.E. Flue Tube Recuperators for Reheating Furnaces



One of two M.E. Flue Tube Recuperators supplied to Stein & Atkinson Ltd., for two oil fired slab reheating furnaces, each with a capacity of 60-70 tons per hour, built by them for the Brinsworth Medium Strip Mill of Steel, Peech & Tozer Ltd. Each recuperator heats 1,140,000 c.f.h. of combustion air to 400° C.

M.E. FLUE TUBE RECUPERATORS HAVE THE FOLLOWING ADVANTAGES:—

1 They are supplied as an easily erected single unit of all steel welded construction, pressure tested in our Works to twice the working pressure.

2 The flue gases pass through plain bore seamless tubing which can easily be cleaned.

3 They are compact in size, taking up little floor area and can be cleared by shop cranes.

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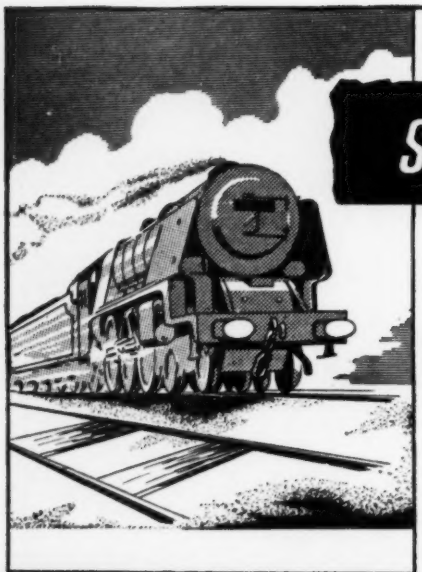
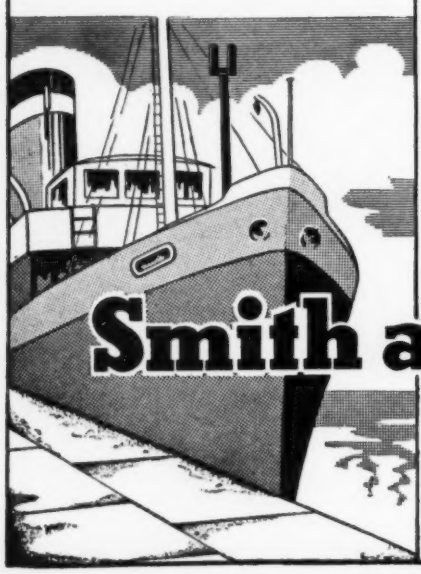
We make many other types of all steel recuperators — please write for descriptive booklet





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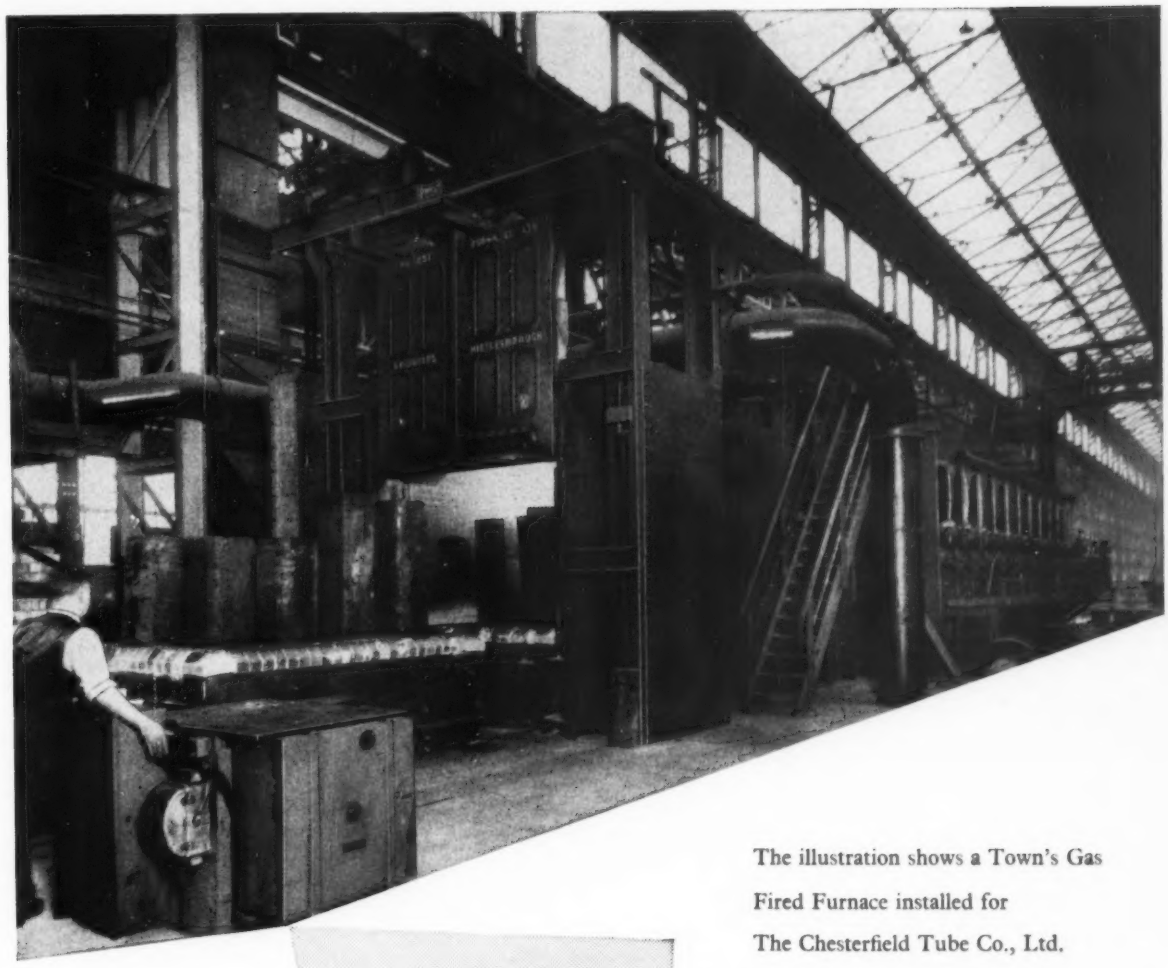

The illustration shows a Town's Gas Fired Furnace installed for The Chesterfield Tube Co., Ltd. This unit is for heating billets prior to piercing and is also suitable for heat treating miscellaneous items in the soaking section.

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We specialise in the design and construction of:

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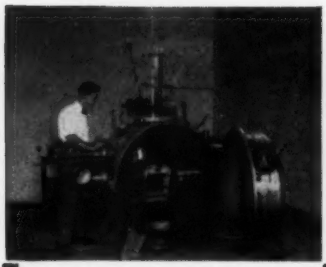
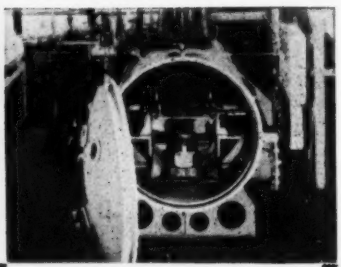


Model 2555 Vacuum Induction Furnace with melting capacity of 50 pounds of steel. Other standard furnaces have capacities of 12 to 3,000 pounds.



Model 2705 Non-Consumable Arc Skull Furnace with a capacity of 50 pounds of titanium. Other standard vacuum arc furnaces have capacities of 8 to 10,000 pounds of titanium.

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for* all *heat-treatment purposes

Metals for parts to withstand jet engine temperatures and stresses can be produced in vacuum furnaces. Such metals enable components to be made with fewer rejects. If your products need parts that operate under severe conditions or need high purity metals or alloys—a vacuum furnace will help you to produce them.

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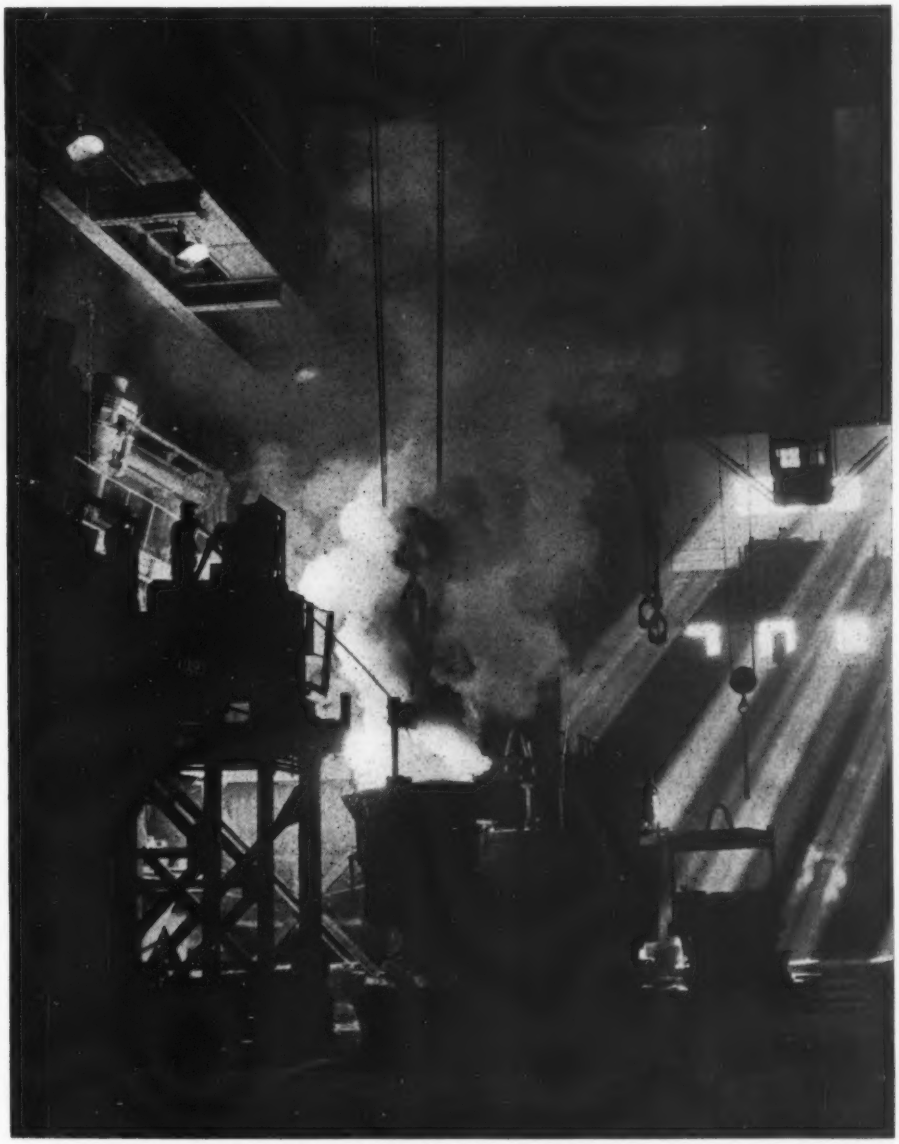

Furnace tapping in the Electric Arc Melting Shop of Messrs Thos Firth & John Brown Ltd Sheffield, by whose courtesy this photograph is reproduced

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Mine cars made from SCW Cor-Ten (Photograph by courtesy of the National Coal Board)

Where advantage can be taken of its high yield strength and corrosion resistance, SCW Cor-Ten can always effect a substantial cost saving.

Whatever the application (and the range continues to grow) SCW Cor-Ten is able to offer an unchallengeable combination of economy and efficiency.

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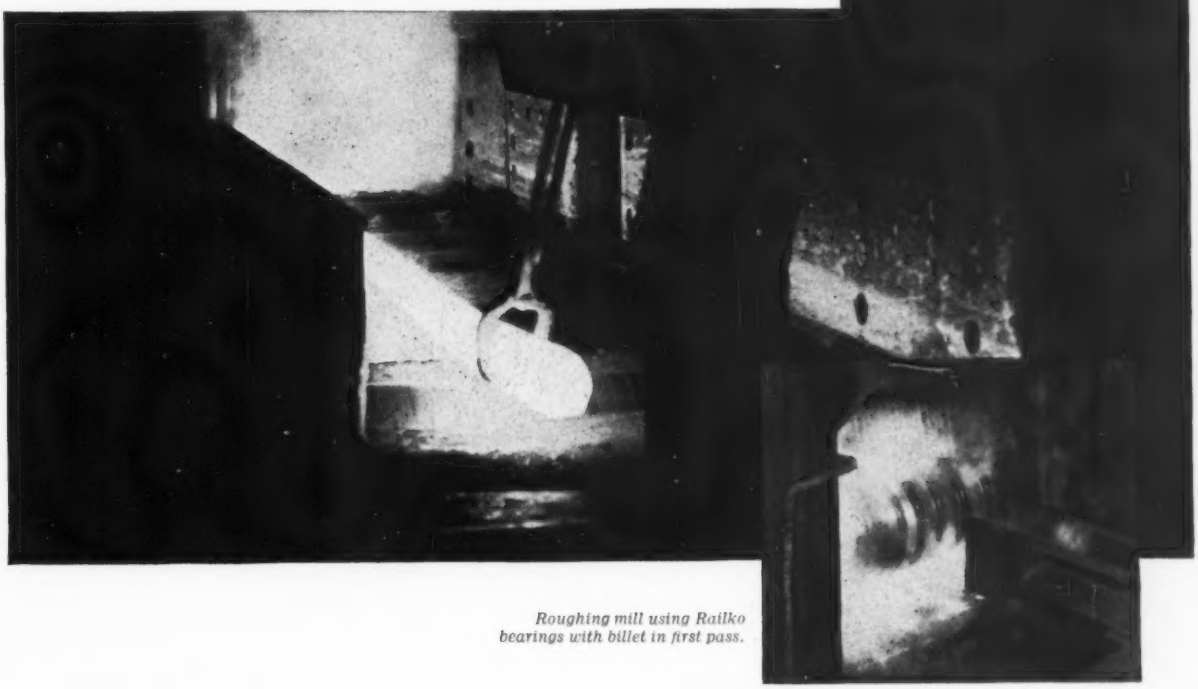
Railko Bearings at Skinningrove—

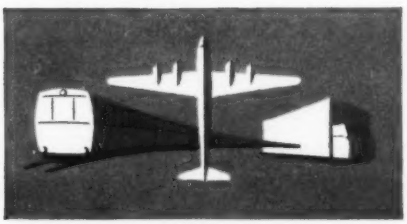


Roughing mill using Railko bearings with billet in first pass.

TOUGH JOB—TOUGH BEARINGS . . .

At Skinningrove, as at many other leading steel works, Railko bearings have proved far more satisfactory where operating conditions are difficult than bearings of conventional materials.

Why do Railko bearings do a better job? . . . they are tough resin-impregnated fabric laminates specially moulded under heat and pressure to give specific properties depending on the work involved. Railko bearings have many advantages. They withstand extreme pressure and movement * have a high strength-to-weight ratio a high compressive strength * a low co-efficient of friction * can be lubricated with a variety of fluids * in certain applications can be used without fluids are resistant to acids and alkalis * have a low rate of wear * are easily machined have excellent weathering properties.



Railko bearings are now widely used by manufacturers in the steel, railway, aviation, engineering and chemical industries. Perhaps they can help *you?* Write to us and we shall be glad to send you full technical details.

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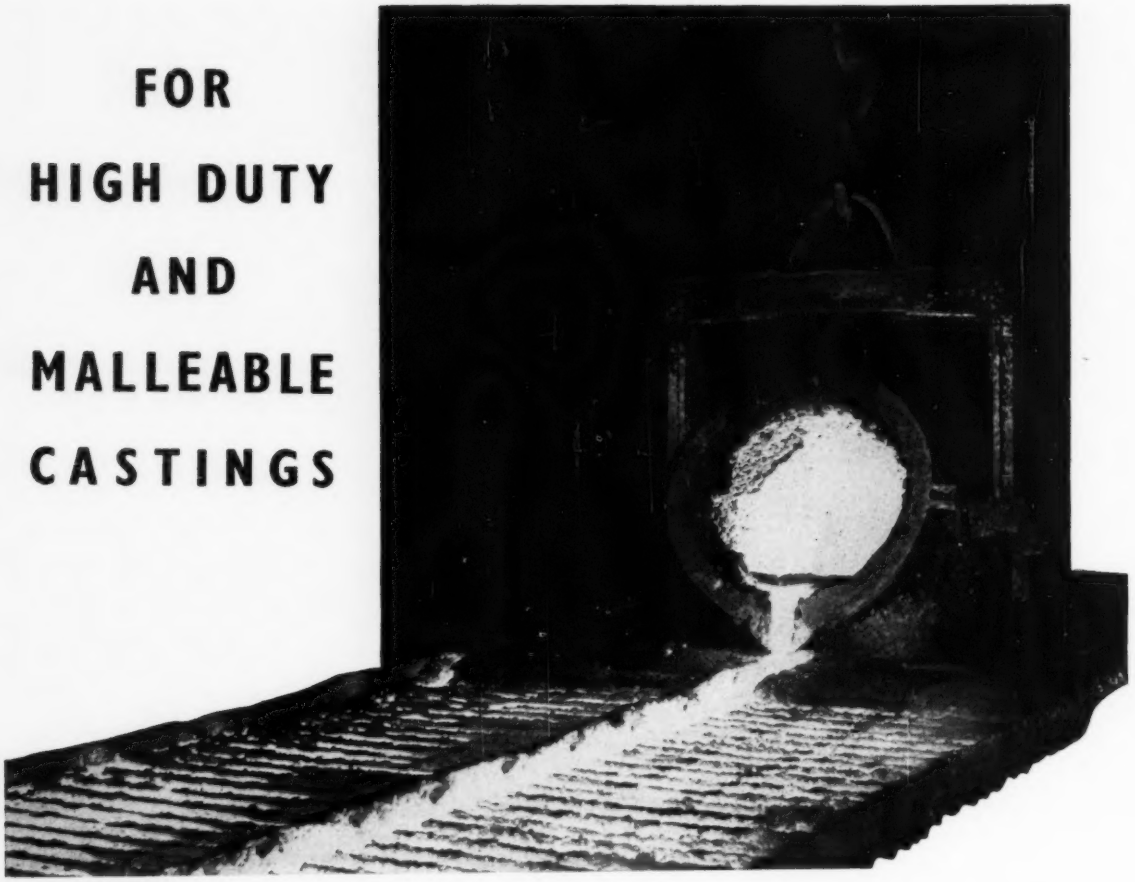

The seven standard grades of Stanton Dale Refined Pig Iron

GRADES:	A	B	C	D	E	F	G
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PHOSPHORUS %	0·6–0·7	0·6–0·7	0·6–0·7	0·4–0·5	0·25	0·25	0·25
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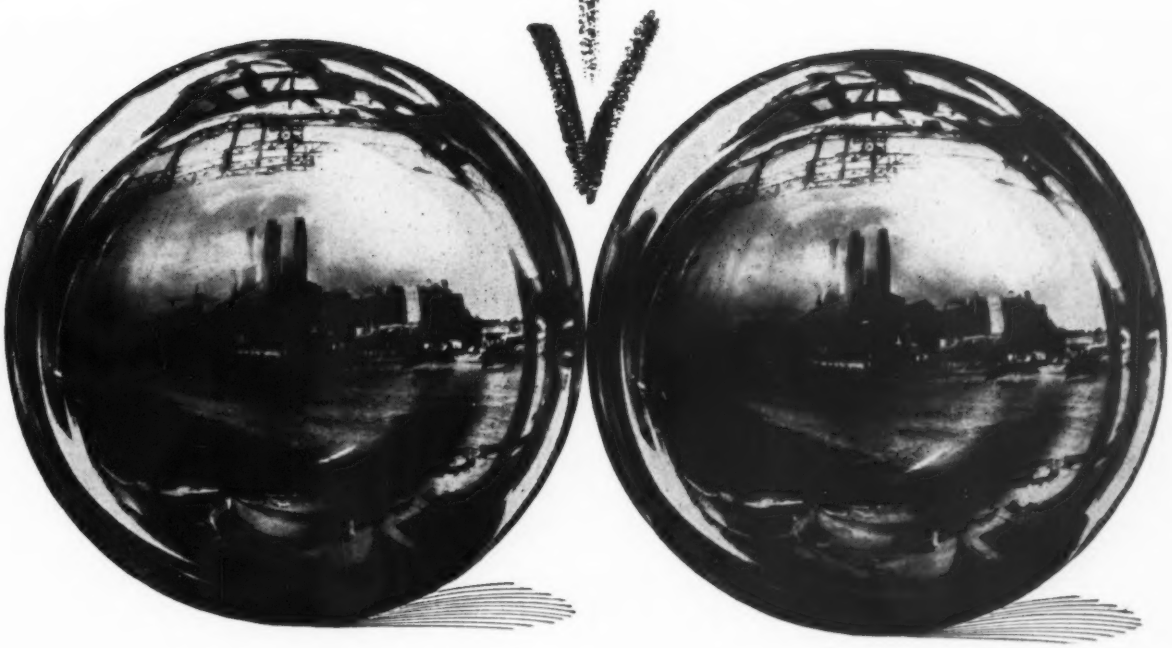

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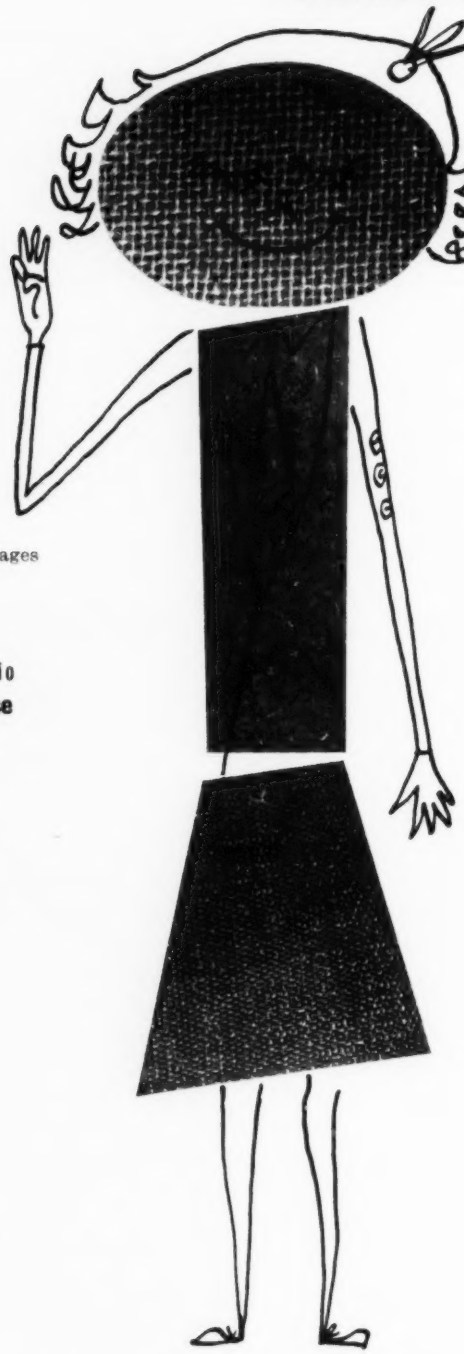

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Journal of The Iron and Steel Institute

Contents

THE IRON AND STEEL INSTITUTE

A review of steelmaking processes *by T. P. Colclough* 201
The effect of copper on martensite breakdown and hardenability of a 0·3%C steel *by A. S. Kenneford* 215
Controlled-transformation stainless steels *by K. J. Irvine, D. T. Llewellyn, and F. B. Pickering* 218
The stress-dependent damping capacity of ferromagnetic metals *by G. Sumner and K. M. Entwistle* 238
An electrolytic jet machining technique for the production of thin foils of steel *by P. M. Kelly and J. Nutting* 246

BRITISH IRON AND STEEL RESEARCH ASSOCIATION

The mechanical properties of hardened high-speed tool steels – a review of methods of determination *by J. Clayton-Cave and E. Ineson* 248
The spectrographic determination of residual elements in steel – the excitation index technique *by the BISRA Spectrographic Analysis Sub-Committee* 253
The tempering of 2¼%Cr – 1%Mo steel after quenching and normalizing *by R. G. Baker and J. Nutting* 257

LETTER TO THE EDITOR

Austenite transformation and the Greniger–Troiano technique *by E. G. Ramachandran and C. Dasarathy* 268

DISCUSSION ON PAPERS

Discussion at the Autumn General Meeting, 1958, on: continuous casting 269
creep in metals 280

NEWS 298 **ABSTRACTS** 301

TRANSLATIONS 300

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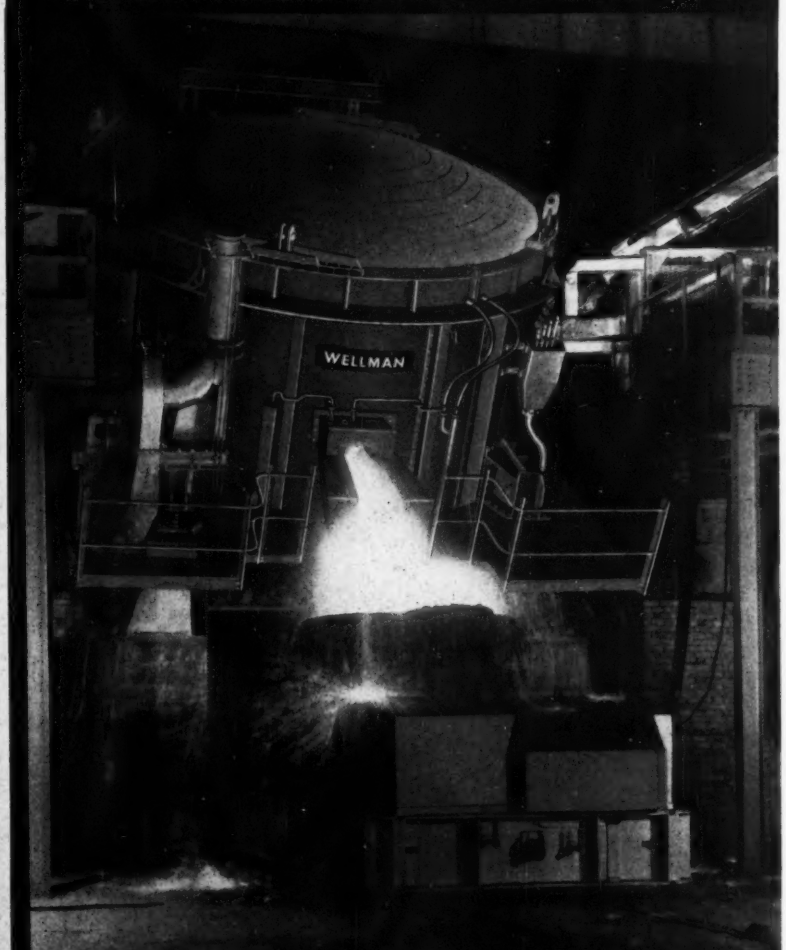

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SIR JULIAN PODE, J.P.

SIR JULIAN PODE was born in Sheffield on 26th June, 1902. He was educated in HMS Conway, and served in the Royal Navy during the First World War.

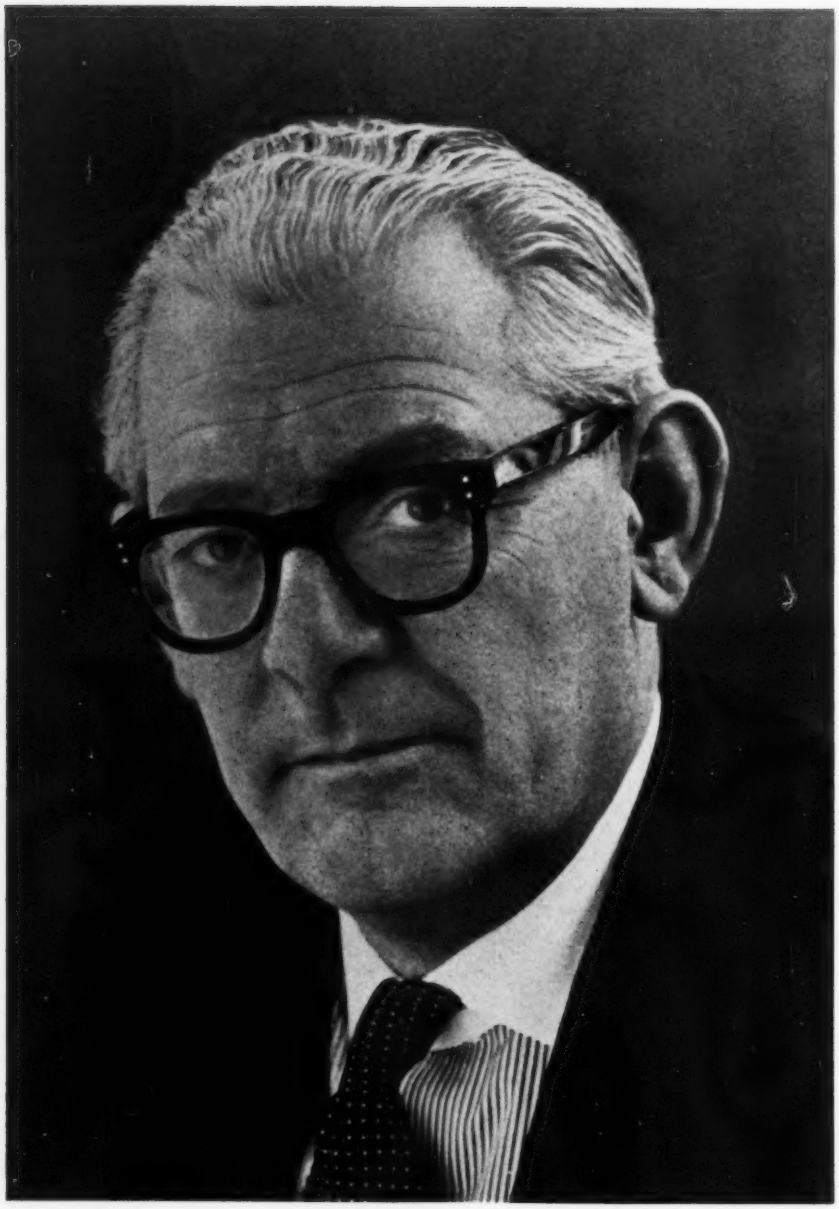
In 1926 he joined the Dowlais works of Guest Keen and Nettlefolds Ltd as accountant, later becoming district accountant. When GKN and Baldwins Ltd amalgamated their heavy steel interests in 1930, Sir Julian was appointed secretary of the new group, Guest Keen Baldwins Iron and Steel Company Ltd, and he was closely associated with the rebuilding and modernization of the works at East Moors, Cardiff, in 1936.

Two years later, with Mr C. R. Wheeler, C.B.E., Sir Julian became joint commercial manager of Guest Keen Baldwins Iron and Steel Company Ltd, while remaining secretary of the company. He then worked on plans for modernizing the steel sheet and tinplate industry in South Wales, which were finally carried out in 1947 when a new company, The Steel Company of Wales Ltd, was formed with Sir Julian as managing director.

Sir Julian's many interests and activities include membership of the Executive Committee of the British Iron and Steel Federation and a number of the Federation's committees and associations. He is a director of the National Industrial Fuel Efficiency Service, a member of the Grand Council of the Federation of British Industries and of the Industrial Coal Consumers' Council. He is also a director of The Steetley Company Ltd, Benzole Producers Ltd, B.I.S.C. (Ore) Ltd, Guest Keen and Nettlefolds Ltd, and The Iron Trades Employers Insurance Association Ltd. He is deputy chairman of the Development Corporation for Wales, chairman of the Management Committee of the Athlone Fellowship Scheme, and a member of the South Wales Committee of Lloyds Bank Ltd.

Sir Julian Pode was High Sheriff of Glamorgan in 1948 and was appointed a Justice of the Peace in 1951. He was made an Honorary Freeman of the Borough of Port Talbot in 1957, and in the New Year's Honours List of 1959 he was created a Knight Bachelor.

Sir Julian joined The Iron and Steel Institute in 1939 and became a Member of Council in 1952. He was elected Vice-President in 1957 and this year he succeeds Mr W. Barr, O.B.E., as Honorary Treasurer of the Institute.



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A review of steelmaking processes

By T. P. Colclough, C.B.E., D.Sc.

SYNOPSIS

This paper gives a brief review of the traditional methods of steelmaking, a heat balance for each process, and an outline of the factors underlying the selection of the method employed for different local conditions.

The application of tonnage oxygen has created a revolution in the making of steel by the pneumatic process. It is now possible to make Thomas steel of low nitrogen content and to utilize pig iron of low and medium phosphorus content in a basic converter. This application makes it possible to produce commercial steels with the known economic advantages of the pneumatic process equal to or approaching OH quality.

An examination is made of the application of tonnage oxygen to the pre-refining of hot metal, enrichment of the combustion air, and oxygen lancing of the liquid metal in OH furnaces. With these developments in technique, the OH process will maintain its advantages as regards quality of product and flexibility of operation and become more competitive on the economic basis. 1728

THE RAPID DEVELOPMENTS which have been made in the techniques of steelmaking in the post-war period naturally give rise to the question as to which process offers the optimum attraction in any particular set of conditions. It is therefore desirable to examine the principles on which the different modifications are based, to consider their application to the available raw materials and, most important, to consider the effect of possible or probable further developments.

It may be accepted that perhaps the simplest approach to the study of the principles involved in the new techniques may be found in a rapid survey of the steelmaking processes hitherto practised. In making this review, attention is deliberately confined to the large-scale manufacture of mild commercial qualities and does not extend to the highly specialized techniques involved in the production of high-quality carbon and alloy steels on which the reputation of UK steelmakers has been rightly based.

In essence, the steelmaking process comprises the elimination of some 6 or 7% of impurities from the pig iron used, raising the temperature of the liquid metal to the level necessary for the clean casting of ingots, and the formation of a fluid slag which will carry off the impurities and protect the surface of the liquid steel. In all processes, this removal of the impurities is effected by the use of oxygen, either in gaseous form or as solid iron oxide, and the heat required is furnished either internally from the oxidation process or by external fuel.

Paper presented to the Annual General Meeting of the Institute, May 1959.

Dr Colclough is technical adviser to the British Iron and Steel Federation.

The details of the operations in the three principal processes, acid Bessemer, basic Bessemer, and open hearth, are too well known to require elaboration but certain important aspects merit review.

ACID BESSEMER PROCESS

In the acid Bessemer process, liquid pig iron is blown with atmospheric air. The time taken for the removal of the impurities (C, Si, Mn) is of the order of 15 min and the complete operation, tap-to-tap, occupies 35–40 min. No external fuel is applied and the whole of the heat requirement is furnished by the oxidation of these elements, together with that of some metallic iron. A fluid slag is formed by the combination of the silica with FeO and MnO formed in the oxidizing process.

A brief outline of the materials, oxidation, and thermal data is given in appendix A and summarized in Table I. This table shows that with careful operation and 'good housekeeping' the yield of ingot of

TABLE I
Acid Bessemer process

MATERIALS BALANCE per ton of mild steel			
Input	lb	*Output*	lb
Hot metal	2505	Oxidized metal	268·5
		,, additions	4·0
Return scrap	45	Ejections (0·5%)	12·5
Additions	20	Casting scrap	45·0
		Ingots	2240·0
	2570		2570·0

HEAT BALANCE 10^3 Btu					
Available		*Used*			
					%
Liquid metal	1187·3	Heat in: steel		1275·7	49·1
Scrap	..	slag and fume		225·5	8·7
Oxidation: C, Si, Mn, Fe	1393·1	Gases (1720° C) N_2 10230 ft³		685·4	26·4
Slag formation	15·6	CO 3320 ft³		224·7	8·7
		Radiation and losses		184·7	7·1
	2596·0			2596·0	100·0

Heat added to metal, slag, and fume 22·2% of heat generated
Yield of ingot 87·1% of metallics charged

TABLE II
Basic Bessemer process

MATERIALS BALANCE per ton of mild steel			
Charged	lb	*Product*	lb
Hot metal	2500	Oxidized metal	263
		,, additions	4
Return scrap	45	Ejections (0·5%)	13
Additions	20	Casting scrap	45
		Ingots	2240
	2565		2565

HEAT BALANCE 10^3 Btu					
Available		*Used*			
					%
Liquid metal	1185·0	Heat in: liquid steel		1275·2	47·0
Scrap	..	slag and fume		456·7	16·8
Oxidation: C, Si, Mn, P, Fe	1299·6	Gases (1720° C) N_2 9736 ft³		652·3	24·0
Slag formation	228·7	CO 2800 ft³		189·6	7·0
		Radiation and losses		139·5	5·1
	2713·3			2713·3	99·9

Heat added to metal, slag, and fume 36% of heat generated
Yield of ingots 87·3% of metallics charged

mild steel is about 87·0% of the metal and additions charged. In comparing yields, it will be assumed in every case that good housekeeping conditions are established. In practice, the yield obtained may be some 1% lower than will be quoted in this paper.

The total heat involved in the operation is about 26 therms/ton of ingot, of which 12 therms are supplied in the initial heat of the liquid iron and 14 therms by the oxidation process. The operation, thermally, is of low efficiency, the heat added to the metal and slag representing only about 22% of the heat which is generated, and 64% of the heat generated escapes in the issuing gases passing up the chimney. This efficiency is even lower still if regard is paid to the fact that the carbon in the metal is oxidized to CO only. If this CO were oxidized to CO_2, a further 10 therms would be released. It is also to be noted that the margin of heat available is low and there is no surplus heat for the melting of scrap additions or other materials.

The main handicap of the process arises from the fact that it is not possible with an acid lining of the converter to remove either S or P from the metal and none but pure metal can be used. With the high cost and scarcity of low S and P ores, the low yield of ingot, and the limitations imposed by the absorbed N_2, it is clear that there is no future in this air-blown process, except in certain special conditions.

BASIC BESSEMER PROCESS

The discovery by Sidney Gilchrist Thomas that the converter could be built with a dolomite lining opened up an entirely new field of operations, for then both S and P could be removed from the liquid iron. The use of a basic lining demands that the Si content of the pig iron shall be as low as is practical and that lime must be added to combine with the SiO_2 and P_2O_5 to form slag. This addition of CaO and the increase in slag weight leads to an increase in thermal requirement and to meet this the normal P content of the iron used is about 1·8–2·0%.

An outline of the materials and thermal data for a typical basic Bessemer charge is given in appendix B and summarized in Table II. This table shows that the thermal requirements per ton of ingot are 27 therms and the yield of ingots is 87·3% from the metallics charged, or roughly the same as in the acid process. The operating cost of the basic process is higher than that of the acid converter because burnt lime must be added, but this is more than offset by the value of the slag formed. Using pig iron of roughly 2%P and low Si, the slag formed is about 25% of the ingot weight and contains about 18%P_2O_5. This slag has a ready sale as fertilizer and the credit arising is an important part of the economy of the basic Bessemer process.

It is to be noted that, as in the acid Bessemer process there is no surplus heat available for the melting of other than a limited quantity of scrap.

In both these processes, the operations are practically continuous, two or three vessels being installed to maintain continuous production. The rate of production is extremely high. With this practically continuous operation (a tap every 35 min), there is a high rate of utilization of manpower and relatively low labour costs. Operating costs are therefore low.

The plant required is simple in construction, relatively low in capital cost and with the high productivity this gives low capital charges. On the other hand, the very rapidity of the process presents difficulties in control by laboratory or other testing methods, and to some extent the quality of the product is liable to variability. In addition, there is the most important factor that the steel made by both the Bessemer processes absorbs N_2 from the air blown. This absorbed N_2 is deleterious in commercial steels of low carbon content and as a result, the use of the basic Bessemer or Thomas process has been largely restricted to the manufacture of steel for common purposes. For this steel there has been a large market and it has

TABLE III

OH furnace process, low pig-iron charge, cold metal (producer-gas fired)

MATERIALS BALANCE, per ton of ingot

Charged	lb	*Product*	lb
Cold pig iron	735	Oxidized metal	134
Clean scrap	1690	,, additions	4
Additions	20	Tapping loss	11
		Casting scrap	56
		Ingots	2240
	2445		2445

HEAT BALANCE 10^3 Btu

Available		*Used*		%
In metal charged	nil	Heat in: metal	1287·5	
Oxidation of metal	832·6	slag and fume	284·1	
Slag formation	81·0		1571·6	18·2
Heat in regeneration (1200° C) air	1946·0	Decomposition of $CaCO_3$	81·0	0·9
gas	1273·0	Absorbed in regeneration	3219·0	37·3
Producer gas combustion	4500·0	Chimney gases at 750° C	2154·0	24·9
		Radiation and other losses	1607·0	18·6
	8632·6		8632·6	99·9

Useful heat in metal, slag, and fume 30% of heat generated
Yield of ingots 91·6% of total metallics charged

TABLE IV

OH furnace process, 65% hot metal

MATERIALS BALANCE, per ton of ingot

Charged	lb	*Product*	lb
Hot metal	1500	Oxidized metal	179
Clean scrap	850	,, additions	4
Oxides Metal (200 lb)	120	Tapping loss	11
Additions	20	Casting scrap	56
		Ingots	2240
	2490		2490

HEAT BALANCE 10^3 Btu

Available		*Used*		%
Hot metal	711	Heat in: liquid steel	1357	15·6
Scrap	—	slag and fume	353	4·0
Oxidation of metal	1331	Decomposition of lime and oxides	498	5·7
Slag formation	93	Absorbed in regeneration	2245	25·8
Regenerated air	2245	Chimney gases	1854	21·2
Fuel and steam	4341	Radiation and losses	2414	27·7
	8721		8721	100·0

Heat absorbed in metal, slag, and fumes 26% of heat supplied and generated

Ingot yield: 95·3% of metal charged
90·0% of total metallics charged

sold at a lower price than that of the higher grades of steel.

As indicated, neither of these two processes with air-blowing has any margin of heat which can be used for the melting of scrap, and an alternative method of steelmaking is obviously imperative. In consequence, the OH furnace, which was originally a scrap-melting operation, was developed.

OH PROCESS

Cold metal

In contradistinction to the Bessemer processes, the raw materials used in the early periods of OH development consisted of cold pig iron and scrap, and the operation included the melting of the materials as well as the refinement of the metal. In common practice, the charge consisted of about 30% of pig iron and 70% of scrap.

Some details of a typical charge are shown in appendix C and summarized in Table III. The fuel used was producer gas which, in the early days, was the only available fuel. Since only 30% of the charge is pig iron, a very large part of the oxidation work is carried out by the excess air present in the furnace and in the furnace gases. The table shows that the ingot yield from the metal charged is about 92·4% or 91·6% from the total metallics charged.

It is to be noted that the heat requirement is of the order of 45 therms and the heat in the metal and slag represents about 30% of the heat generated by oxidation and supplied as fuel. In comparison with the Bessemer processes, the operation in the OH furnace is definitely slow, some hours being required for each charge and, although the CO generated by the oxidation process is burnt to CO_2 instead of escaping as CO, as in the Bessemer processes, the amount of heat which is lost by radiation and other factors is about 20% of the total heat in the system, or over 30% of the heat generated.

Compared with the Bessemer methods, the OH scrap melting operation is definitely slow, gives a slightly higher yield, and demands a higher fuel consumption. On the other hand, the very slowness of the operation makes control relatively simple and certain and the quality of the steel made is superior by virtue of this better control and its much lower content of deleterious nitrogen.

Hot metal

Apart from Western Europe and Sweden, the main bulk of the iron ores available gives a pig iron with a P content intermediate between that of the acid and the basic Bessemer qualities. It therefore became necessary to develop processes which would successfully handle both scrap and this type of pig iron. With the growth of the industry, it became normal to integrate coke, pig iron, and steel manufacture whereby the pig iron is transferred to the steelmaking furnaces in the liquid state. As a result, the modern OH furnace deals with liquid pig iron of different P content and varying proportions of scrap.

The materials, and oxidation and thermal data for a typical hot-metal charge made up of about 65% hot metal and 35% scrap are given in appendix D and summarized in Table IV. Using mixtures in this proportion, the amount of oxidation work to be carried out is intermediate between that of the cold metal process and the Bessemer process. This oxidation is carried out partly by the furnace gas and partly

TABLE V
Steelmaking balance in Germany

Scrap available: 40% of ingot weight			
Process	Ingots made	Materials required	
		Pig iron	Scrap
OH	100	33	77
Basic Bessemer	100	112	3
Total	200	145	80
Average	100	72·5%	40%

by the addition of iron ore. The reduction of iron oxide absorbs heat but, on the other hand, adds metallic iron to the bath, replacing part of the metalloids oxidized. As a result, the yield of ingot from metal and scrap by this process averages a little over 95% of the metal charged, about 7% better than the Bessemer yield, but taking into account the Fe reduced from the iron oxide the overall yield is about 90% of the total metallics charged, or about 3% higher than in the basic Bessemer process.

As in the cold metal process, the hot metal process is relatively slow and the radiation and other losses are excessive. The furnaces are, in modern practice, fired by fuel oil and the fuel used in good practice averages about 43 therms/ton. The useful heat added to the liquid metal and slag and absorbed in the decomposition of the fluxes and oxides amounts to only 26% of the heat supplied and generated. About 34% of the heat so generated passes away in the chimney gases, and while a substantial part of this can be recovered by the use of waste-heat boilers, about 40% of the heat generated is lost by radiation and other factors.

The operating cost of the OH furnace is substantially higher than that of the Bessemer converter. This arises from the necessity to apply external fuel, coupled with the low thermal efficiency of the process and the lower productivity of the manpower employed. This differential is modified to some extent by the higher yield and by the lower cost of the scrap used than of the pig iron which it replaces.

Similarly, the capital cost of an OH shop is considerably higher than that of a Bessemer converter plant of the same annual production, thereby giving a higher capital charge per ton of ingot. This simple statement, however, requires qualification. In an integrated steelworks there must be a balance of the iron- and cokemaking plant with the steelmaking. For each 100 tons of ingots, the Bessemer plant requires about 110 tons of hot metal, as compared with, say, 60–65 tons for the OH process. Since in modern plant the capital cost per ton of ingot for blast furnaces and coke ovens is about 2½ times that of the steelmaking plant per ton of ingot, it follows that the overall capital charges per ton of ingot, up to the ingot stage, are roughly 20% higher for the Bessemer plant than for the OH process.

No general comparison can be made because of local variations but there can be no doubt that the total cost of basic Bessemer steel is lower than that of OH steel and this must be taken care of by quality and other considerations.

Selection of steelmaking processes

The selection of the correct process to adopt in different sets of conditions sometimes presented considerable difficulties. In countries like Belgium, Luxembourg, and Eastern France, the selection was simple. The local ores, mined at low cost, gave pig iron of roughly 2%P and the choice of the basic Bessemer process was an obvious one. A large proportion of the steel made was of ordinary commercial quality and was rolled into semi-finished steel or products which gave a relatively high yield. As a result, the quantity of scrap arising in the steelworks was very limited and the amount available could be readily dealt with by a few OH furnaces, electric furnaces, or by export. A very large proportion of the Thomas steel made in Belgium and France is for local consumption, while that for Luxembourg is largely exported for common purposes.

In Sweden, where both high-purity and high-phosphorus content ores are available, either the acid or basic Bessemer process can be adopted, subject to the limitation of the high N_2 content of the low-carbon steels.

In Italy, where ore and pig-iron supplies were strictly limited, the only readily available material was scrap and therefore the steelmaking process adopted was the cold-metal OH method or the electric furnace.

On the other hand, in Germany, where the steelmaking and engineering industries are closely associated, the amount of scrap arising within the country is of the same order as in the UK. Experience shows that the amount of scrap arising from industry and general use in a highly industrialized country is roughly 40% of the weight of steel ingots made. Most of the domestic ores in Germany are of the phosphoric type, and with high-phosphorus ores available for importation from Sweden, about 50% of the steel in Germany was made by the basic Bessemer or Thomas process. The other 50% of ingot production was, broadly speaking, made by the cold-metal OH process, using a charge of 70% scrap and 30% low phosphorus manganiferous pig iron known as 'stahleisen.' Assuming that the basic Bessemer process uses 112 tons of hot metal and 3 tons of scrap, and that the cold metal OH process uses 33 tons of pig iron to 77 tons of scrap for each 100 tons of ingots, it will be seen from Table V that this 50 : 50 proportion of Bessemer and OH steel gives an internal balance in scrap consumption and supplies. This arrangement proved very satisfactory, the basic Bessemer steel meeting the requirements of the common steel market, while the OH furnaces gave a substantial proportion of high-quality steel to meet the requirements for which Bessemer quality was not suitable.

In the USA, where the general quality of the ores gave a pig iron with about 0·3%P, the pig iron was not suitable for use in the basic Bessemer process. The reserves of low phosphorus ores suitable for the acid Bessemer process are practically exhausted and over the last 40 or 50 years steel manufacture has been largely confined to the OH and electric furnace processes. The acid-lined converter is still in common

use but today its main function is to pre-refine the ordinary pig iron before charging into the large OH furnaces.

Similar conditions apply in the USSR and this has led to the adoption of the OH process as the standard method of steelmaking.

That, in general, was the position pre-war. What are the changes which have taken place? First, we shall deal with the position in countries using high-phosphorus pig iron.

High-phosphorus pig iron

The experience in Germany and Western Europe during the Second World War accentuated the already well known inferiority of Thomas steel quality arising from the inherent nitrogen content of the blown steel in comparison with OH steel.

The increasing stringency of specifications and particularly the growing demand for deep-drawing steel of low S and P contents, intensified this disparity and it was obvious that steel of the traditional Thomas quality could not be regarded as competitive in spite of its lower cost. At the same time, it had to be recognized that the pig iron made from indigenous ores was not suitable for the OH process, that a very large proportion of the steel made in Western Europe came from the Bessemer plants, and that a very heavy capital expenditure would be incurred if the existing plants had to be replaced by some other steelmaking process. The position was critical and a solution had to be found.

Efforts were made during the war to reduce the N_2 content of Thomas steel by the use of the HPN process, and other modifications, with some, but definitely limited, success. It was recognized that the only reliable solution was the elimination of air-blowing and its replacement by oxygen. The advent of tonnage oxygen at an economic price opened up a completely new range of possibilities and the developments in different countries have naturally varied according to local conditions.

In Western Europe, the main tendency has been to maintain the existing Bessemer plants and to apply oxygen to the traditional bottom-blown practice. In the earlier developments, oxygen was added to the blast to enrich the oxygen content. This reduced the actual volume of wind blown and also the nitrogen percentage in that wind. This method naturally gave some improvement as regards N_2 content but cannot be regarded as entirely satisfactory and has led to further development.

It is well known that with air-blowing, the nitrogen content of the metal falls to a minimum value of about 0·003% in the early stages of the carbon blow and thereafter increases fairly rapidly. In the modern practice, air is blown, if at all, for 3 or 4 min and is then replaced by an oxygen blast using the traditional basic tuyere block. The use of pure O_2 as blast naturally gives rise to extremely high local temperatures at the mouth of the tuyeres with consequent excessive wear. This difficulty has been overcome by modification of the design and construction of the tuyere block.

In addition, the use of pure O_2 and the absence of the 80% of N_2 in the blast leads to excessive general temperature conditions in the metal bath, and thus excessive wear on the converter lining and overheating of the metal, especially in the after-blow period. *It is essential to employ a cooling agent and this excessive heat has been counter-balanced by the use of steam or* CO_2 in the tonnage O_2 blast. This steam (or CO_2) maintains a physical flow of gas in the metal at all stages of the blow and by its decomposition furnishes oxygen and absorbs heat, thereby restoring thermal conditions similar to those of the normal air blow. In practice, the steam used gives a blast of roughly 50%O_2 and 50% steam. The thermal conditions are summarized in Table VI.

It is to be noted that, using the same quality of hot metal there can be little or no change in the fundamentals of the process. The yield of ingots and the weight of slag are unaltered; the time of blowing shows a slight reduction owing to the smaller volume of the O_2–steam blast, and there is a similar reduction in the volume of the waste gases. The total heat requirement per charge of 27 therms is practically unaltered, but the amount of heat carried away in the exit gases shows a drastic reduction, the difference being the endothermic heat absorbed in the metal by the decomposition of the steam.

There is a natural increase in operating cost arising from the cost of the tonnage O_2 (about 1725 ft^3/ton) and steam (82 lb) consumed.

The fundamental factors, however, are that steel of low N_2 content is now made from the normal pig iron in the established Thomas plants with the same high production rates as previously and without substantial capital expenditure other than that of the oxygen plant. The merits of the process may be judged from the fact that practically the whole of the Thomas plants in Western Germany, France, and Luxembourg have been converted to oxygen–steam blowing. In Belgium, O_2–CO_2 blowing has been developed with the same purpose in view. In the UK the O_2–steam blast has been adopted at the two existing basic Bessemer plants and this process is the basis of the new plant under construction at The Steel Company of Wales.

Low-phosphorus pig iron

The problem presented in Austria was entirely different in character and the solution evolved merits study because the principles entailed are capable of wider application.

The supply of scrap in Austria is strictly limited and the steelmaking process must be based on a high proportion of liquid pig iron. The iron made from the local ores is of somewhat unusual analysis, containing about 3%Mn with 0·2%P.

Iron of this low P content is not suitable for tilting OH practice because there is no fertilizer slag credit and it cannot be blown in the basic Bessemer converter because of the lack of heat generated. For comparative purposes, typical data for a hypothetical basic Bessemer blow using 0·3%P hot metal are given in Table VI. It will be noted that while the materials, oxidation, slag formation, etc. (details of which are given in appendix F) are naturally parallel to those of the normal converter operation, the heat carried away by the metal, slag, and waste gases

TABLE VI

Basic Bessemer process, comparison of air and O_2–steam blowing

HEAT BALANCE 10^3 Btu				
Available		*Used*	Air blown	O_2–steam
Hot metal	1185	Heat in steel and slag, etc.	1732	1732
Scrap		Sensible heat in gases N_2	652	nil
Oxidation	1300	CO	190	190
Slag formation	229	H_2	nil	110
		Decomposition of steam	nil	475
		Radiation, etc.	140	207
	2714		2714	2714

TABLE VII

Basic Bessemer process, low phosphorus iron (theoretical estimate)

MATERIALS BALANCE, per ton of ingot			
Charged	lb	*Product*	lb
Hot metal	2443	Oxidized metal	207
Scrap	45	additions	4
Additions	20	Ejections	12
		Casting scrap	45
		Ingots	2240
	2508		2508
HEAT BALANCE 10^3 Btu			
Available		*Used*	
Hot metal	1158	Heat in: steel	1275
Scrap		slag and fume	246
Oxidation	967	Gases (1720° C)	
Slag formation	62	N_2 8184 ft³	548
	2187	CO 3197 ft³	216
Deficit 10%	223	Radiation, etc. (5%)	125
	2410		2410

Yield of ingots: 90·0% of hot metal and scrap
89·3% of total metallics

at normal operating temperatures with air-blowing is higher by some 10% than the heat available. In other words, the metal will freeze in the converter and it is not practicable to blow metal of this composition in an air-blown converter. It is, however, important to note that with air-blowing, over half of the heat generated is carried away in the sensible heat of the issuing N_2 gas. If, therefore, the air blast is replaced by oxygen, the deleterious effect of the N_2 is removed and the heat shortage is converted into a substantial surplus and blowing of this type of pig iron becomes practicable.

To increase steel production and accelerate the operation of the OH furnaces at Linz and Donawitz, experiments were made to reduce the Si and C contents of the hot metal by blowing the metal with oxygen in a ladle before charging into the OH furnace. Naturally the experiments were successful, and it was decided to continue the blowing operation to the point of steelmaking.

During their studies on the different processes for iron and steelmaking before the war, the Brassert Company investigated the possible applications of oxygen to both the blast furnace and Bessemer converters and, in 1939, a patent was filed by Dr Karl Schwarz and the company to cover the blowing of liquid metal by a jet of oxygen travelling at supersonic speed. If a jet of any fluid, gas or liquid, is travelling at supersonic speed, it has the physical properties of a solid rod. If, therefore, a jet of O_2 is directed at the surface of liquid metal, the O_2 stream will penetrate the layer of covering slag and pass direct into the metal.

LD process

This principle has been adopted and developed by the two Austrian steel companies in the new LD process in which the liquid hot metal is blown in a basic-lined converter with a solid bottom, with the oxygen jet blown through a lance vertically over the centre of the surface of the liquid metal. The reaction is rapid and proceeds at the same speed as in the normal converter, and since the absorption of N_2 is eliminated, the steel made is of very low N_2 content.

As in the O_2–steam-blown basic Bessemer operation, the loss of heat carried away by the nitrogen with air-blowing is eliminated and the metal becomes too hot and must be cooled. This cooling operation may be carried out by the addition of either scrap or iron ore, in contradistinction to the steam or CO_2 used in the bottom-blown converters. Theoretically, about 20% of the ingot weight can be absorbed as scrap but, in practice, this quantity is not normally available and good-quality iron ore or sinter is more usually employed.

This use of iron oxide as a 'coolant' presents factors of fundamental importance and merits detailed study. For this study, it can be assumed that the iron oxide is utilized to oxidize carbon in the metal, as in general steelmaking practice. Heat is absorbed for the heating of the ore and the small amount of slag which is formed from the impurities in the ore and in the actual decomposition of the iron oxide. This heat loss is counterbalanced to some degree by the oxidation of carbon by the oxygen released from the ore and, overall, the nett heat effect of the use of ore is an absorption of roughly 2 Btu/lb of ore added.

In addition to the cooling effect, Fe reduced from the ore adds to the yield of metal; oxygen from the ore reduces the volume of oxygen to be blown, with resulting benefit to the general economy. Both theoretically and practically, it can be demonstrated that about 7% of the ingot weight (157 lb) of a typical ore (56% Fe) can be added as the cooling agent. An analysis of the details of a typical operation is given in appendix G and summarized in Table VI.

It will be noted that the total amount of heat required per ton of ingot, the volume of exit gases, and the sensible heat carried away by the gas are lower in the LD process than in the straight or O_2–steam-blown process. The life of the converter lining is roughly the same for either process, but the LD method has the advantage that the yield of metal is higher and the operating costs are lower because of the elimination of the removable base of the converter and therefore the cost of the 'bottom' house.

The nitrogen content of the steel is about 0·003%, the same or lower than OH quality. This application of oxygen has therefore made possible for the first time the use of the converter process, with its high rate of production, low operating costs, and relatively low capital cost, for the manufacture of low-C steel of OH quality from pig iron of low P content. It is not surprising that the process has attracted considerable attention in the USA since, as indicated earlier, the normal phosphorus content of pig iron there is about 0·3%. It will also become increasingly important in the UK, since many of the new ore resources which will be made available for the expanding steel production here will give a pig iron of similar quality.

Claims are also made that this process can be applied to the manufacture of steel from pig iron of any phosphorus content and also for steel of high carbon quality. Both theoretically and practically, this is obviously correct but certain other factors must be given consideration.

The penetrating power of the O_2 jet is limited. As the P content of the hot metal rises, it follows that the weight, and therefore the depth, of the slag layer must rise proportionally. If the P content of the metal exceeds 0·4–0·5%, it becomes necessary to remove the slag at an intermediate stage and to form a new slag to complete the blowing. This presents no practical difficulty with an experienced crew, but while the time occupied (about 10 min) is relatively short, it represents a substantial part of the total time per blow (35 min) and therefore has a marked effect on productivity, operating cost, and capital charges.

These developments have considerably strengthened the competitive position of the pneumatic processes of steelmaking. What developments have been made in the OH process?

OH process developments

One of the most important developments in the OH process in post-war years has been the replacement of producer gas by fuel oil. While marked improvements in productivity have been shown in several cases as a result of this change, it is perhaps significant to note that the use of this rich fuel has not resulted in any marked increase in fuel efficiency. In the symposium of 1938*, fuel consumption at Irlam is quoted at 3·34 cwt of producer coal. Assuming high-quality coal (13 500 Btu/lb) and a hot-gas efficiency of the producer at 80%, this corresponds to a fuel rate of 38 therms per ton, which is lower than today's rate for the best practice with fuel oil. Obviously some examination of this factor is desirable.

A second factor is that greater attention has been paid to the reduction of the metallurgical load in the OH furnace. With increasing production, the load on the active mixers has grown beyond the point at which effective refining of the hot metal is possible and, in general, the function of the active mixer has tended to become that of a storage vessel with facilities to melt scrap unsuitable for charging direct into the OH furnace.

TABLE VIII

LD process, low-phosphorus iron

METAL BALANCE, per ton of mild steel			
Charged	lb	*Product*	lb
Hot metal	2360	Oxidized metal	213
Scrap	45	,, additions	4
Fe from ore	89	Ejections	12
Additions	20	Casting scrap	45
		Ingots	2240
	2514		2514

HEAT BALANCE 10^3 Btu				
Available		*Used*		%
Hot metal	1119	Heat in: liquid steel, slag and fume	1554	71·7
Scrap				
Oxidation	974	Gases: CO 3090 ft³		
Slag formation	76	N_2 10 ft³	210	9·7
		Reduction of Fe_2O_3	284	13·0
		Radiation and losses	121	5·6
	2169		2169	100·0

Heat usefully employed 68% of heat generated
Yield of ingot 93·1% of metal charged
89·1% of total metallics

For good operation, the OH furnace must be supplied with hot metal of known regular quality. This has led to developments in pre-refining the hot metal by treatment with tonnage O_2 between the mixer and the OH furnace. These developments are proceeding satisfactorily and, coupled with the improvement in iron quality to be derived from the modern blast-furnace practice, they will undoubtedly have a marked beneficial effect on the OH furnace productivity, particularly where increased production is required from existing plant.

This pre-refinement raises the temperature of the hot metal and permits the addition of some scrap to the operation but, in general, pre-refining leads to a loss in yield and more study of this aspect is required.

The OH furnace is seriously handicapped by two factors: its low thermal efficiency and its relatively low speed of operation. By what applications of tonnage oxygen can these factors be improved?

Thermal efficiency

It must be accepted that the operating temperature in the furnace chamber must be maintained above 1 600° C during a large part of the operation and consequently the waste gases, products of combustion of the fuel supplied and of the oxidation process, cannot leave the furnace chamber at a lower temperature. The sensible heat carried out of the chamber is the dominating factor in the thermal efficiency of the furnace. It is correct that part of this heat is returned to the furnace chamber by way of regenerated air or fuel gas and a further part by the raising of waste heat steam, but this recovery has a relatively insignificant bearing on the amount of fuel consumed in the bath

* 'Symposium on steelmaking,' *ISI Spec. Rep.* No. 22, 1938.

TABLE IX

Comparison of fuel requirements of OH furnace practice, 65% hot metal with atmospheric and oxygen-enriched air (per ton of ingot)

	Atmospheric air			Enriched air		
Heat input		10^3 Btu			10^3 Btu	
Heat in metal		711			711	
Oxidation of metal		1331			1331	
Slag formation		93			93	
Fuel	228 lb	4310		150 lb	2835	
Steam		31			21	
Regenerated air at 46·3 Btu ft³	48500 ft³	2245		29800 ft³	1378	
		8721			6369	
Heat use			%			%
Heat in: liquid steel (1620° C)		1357	19·6		1357	26·8
slag and fume		353			353	
Decomposition of oxide and flux		498	5·7		498	7·8
Absorbed in regenerators		2245	25·8		1378	21·6
Chimney gases (750° C)	65100 ft³	1854	21·2	38550 ft³	1156	18·2
Radiation and other losses		2414	27·7		1627	25·5
		8721	100·0		6369	99·9

TABLE X

OH practice 65% hot metal, O_2 lancing

MATERIALS BALANCE			
Charged	lb	***Products***	lb
Hot metal	1500	Oxidized metal	181
		,, additions	6
Clean scrap	830	Tapping loss	11
Ore Fe (237 lb)	142	Casting scrap	56
Additions	22	Ingots	2240
	2494		2494

HEAT BALANCE 10^3 Btu				
Available		***Used***		
Hot metal	711	Heat in: liquid steel	1357	22·8
Scrap		slag, etc.	353	5·9
Oxidation	1333			
Slag formation	93	Decomposition of ore and fluxes	496	8·4
Regenerated air (28200 ft³ ar 1200° C)	1306	Absorbed in regeneration	1306	22·0
Oxygen (650 ft³)	...	Chimney gases (34900 ft³ at 750° C)	1047	17·6
Fuel	2500	Radiation, etc.	1384	23·3
	5943		5943	100·0

Heat absorbed in metal, slag, etc. 38% of heat supplied and generated

Ingot yield: 96·1% of metal charged
89·8% of total metallics charged

and on the speed of operation. The only heat available in the furnace chamber for the steelmaking operation is that heat which is not taken away by the gases.

In Table IV, which represents good modern practice with 65% hot metal fired with fuel oil, the total heat supplied as fuel, by regeneration of air and by oxidation of the bath, is given as 80 therms/ton of ingot. Of this, only 15 therms are absorbed by the metal and slag. About 41 therms are carried away by the issuing gases and of this about half is carried away by the inert N_2 gas.

As in converter practice previously discussed, the replacement of combustion air by tonnage O_2 would permit this heavy load of inert N_2 to be either completely eliminated or reduced with corresponding thermal benefit and therefore reduction of the fuel required per ton of steel.

A long series of tests carried out by the Consett Company, under Mr J. Allen, has demonstrated the value of this principle of O_2 addition to the air for combustion in accelerating the rate of production and lowering heat consumption. It has also been fully demonstrated by the published data of the National Steel Corporation of the USA.

An estimate of the benefits to be anticipated is given in Table IX which shows that with the replacement of 25% of the combustion air by tonnage O_2, giving enriched air of about 26%O_2, the fuel consumption should be reduced to about 16 gal (or 150 lb) of fuel oil per ton of ingot.

With this enrichment, the combustion air would be reduced from 48 500 ft³ to 30 000 ft³ and the volume of the waste gases would fall from 65 000 ft³ to 39 000 ft³, which means a reduction of 40% in the heat per ton of ingot carried into the slag pockets and regenerating system.

Similarly, if the temperature level of the furnace system as a whole be regarded as unchanged, the hourly losses by radiation etc. would not be changed but would show a corresponding reduction per ton of ingot made.

The work already carried out has shown that if the same hourly rates of fuel input, of combustion air, of gas velocities and temperature through the regenerator system, and radiation losses are maintained unchanged, the addition of tonnage oxygen to give about 26%O_2 in the combustion air should give a reduction in fuel consumption of about 35%. Conversely, with the same rate of fuel input per hour, the rate of production should increase by some 50% or, allowing for the time occupied in tapping and fettling, there should be an overall increase in production of about 35%.

The amount of oxygen required for this enrichment would be about 1 370 ft³ and the cost of this tonnage oxygen is offset by the saving of 8 gal of fuel oil.

No substantial capital expenditure is involved in the installation of oxygen equipment for the enrichment of combustion air and the increase in production should yield a substantial reduction in overhead and capital charges. General experience indicates that this is the most simple and direct application of tonnage oxygen to secure increased production and lower costs, particularly in existing plants.

Operating speed

The shortcomings of the OH furnace with regard to operation are too well known to require elaboration. The raw materials, pig iron, scrap, ore, and limestone are charged into the hearth, and gas or liquid fuel is burnt in the space above the charge. At a very early stage, liquid metal is added and the whole

mass of material is covered as quickly as is practical by a layer of slag. Except in the limited early stages of charging, the conditions for heat transfer and rapid reaction are extremely poor.

The heat required must pass from the flame to the metal through the slag, which is a good insulating material. Purification of the metal is effected by oxygen, either from oxide in the slag or by oxygen from the flame passing through the slag. The reaction zone is limited to a shallow area at the interface of the slag and the metal. The impurities in the metal must travel to the interface by diffusion and the oxygen at the interface must be replaced by diffusion in the slag layer.

Further, the normal oxidation by Fe_2O_3 in the metal is endothermic; the metal is cooled, and time and heat are needed to restore the required temperature conditions. These are all time factors. The reactions which take place in 12–20 min in the converter process take hours in the OH furnace, not because of any chemical basis but solely because of physical conditions.

It was demonstrated by the present author more than 30 years ago that, given the desired temperature and slag conditions, raw hot metal could be converted into steel in the OH furnace in 20 min. The only factor determining the length of time of the heat is the slow rate of heat and oxygen transmission through the slag and the relative lack of movement in the metal itself. The essentials for improvement are physical agitation of the metal and generation of heat *within the metal itself*. The advent of tonnage oxygen applied direct in the metal by lancing provides the obvious solution.

Since the oxidation of Si and P by Fe_2O_3 gives a rough thermal balance and no gas is formed, attention can be concentrated on the C/O_2 reaction. If oxygen is blown into the metal at supersonic speed the reaction is vigorous; as in the LD process, carbon is oxidized rapidly and the metal is agitated violently by both the O_2 jet and the gas evolution. The rate of elimination of carbon is limited only, and can be controlled, by the rate of oxygen-blowing. The value of this to secure low-C content of the bath has been fully demonstrated by Dr Kesterton at the Abbey works of The Steel Company of Wales.

Oxidation by O_2 lancing, as in the converter, leads to loss of yield by the loss of carbon and the metal becomes far too hot. As in the converter, *it is necessary to add a cooling agent* and in this case the agent used is iron ore which, for obvious reasons, should be of high quality. The thermal reactions involved should be examined in two stages; in the metal, and in the slag.

Metal

Injecting cold oxygen into a bath of liquid metal at, say, 1 520° C gives for each lb of carbon liberated from Fe_3C and oxidized to CO an evolution of heat in the metal of 5 160 Btu and forms 2·33 lb (31·5 ft³) of CO which will pass through the slag and burn to CO_2. This evolution of heat will obviously raise the temperature of the metal and of the slag.

On the other hand, if Fe_2O_3 is added as a ' coolant ' each lb of carbon oxidized by Fe_2O_3 will require 4·44 lb of oxide and yield 3·11 lb of Fe, increasing the metal weight by 2·11 lb and, again, release 31·5 ft³ of CO. The reduction of the oxide and the heating of the metallic iron reduced will absorb roughly 11 700 Btu/lb of carbon oxidized. This is counterbalanced by the heat of oxidation of C from Fe_3C to CO of 5 160 Btu, giving a nett absorption of heat in the metal of 6 500 Btu/lb of C. This means that the metal is cooled and the reaction slowed down.

TABLE XI

OH practice, 65% hot metal, oxygen-enriched air 750 ft³ O_2, oxygen lancing 650 ft³ O_2

HEAT BALANCE, per ton of ingot 10^6 Btu

Available		*Used*	
Heat in hot metal	711	Heat in: liquid steel	1357
Oxidation	1333	slag (400 lb)	353
Slag formation	93	Decomposition of ore and fluxes	496
	2137		2206
Regenerated air (1200° C) 21700 ft³ at 46·3	1005	To regeneration chambers	1005
Fuel oil (111 lb) = 21 therms	2100	Chimney gas (750° C) 29000 ft³	870
Steam	15	Radiation, etc. (22·3%)	1176
	5257		5257

Heat absorbed in metal and slag: 42% of heat supplied and generated

Slag surface

In both cases the escaping CO will burn with the air in the furnace on the surface of the slag, thereby imparting heat direct to the slag. The combustion of 13·5 ft³ of CO requires 75 ft³ of air, and assuming that the air is regenerated to 1 200° C and the products of combustion (with 20% excess air) escape at 1 720° C, the nett heat evolved on the slag is about 5 300 Btu/lb of C oxidized.

Oxidation of C by Fe_2O_3 cools the metal and tends to slow down the reaction until further O_2 and heat are available. Oxidation by O_2 jet liberates heat in the metal and stimulates physical agitation and reaction. To obtain the best results, the obvious course of action is, as in the converter, to combine the two.

From the above data, each lb of carbon oxidized by Fe_2O_3 absorbs 6 500 Btu while each lb of C oxidized by oxygen jet evolves 5 600 Btu. To balance these actions, 46 lb of C should be oxidized by iron ore and 54 lb by oxygen.

The example of a typical hot-metal practice given in Table IV may be taken as an example for the application of oxygen lancing. Assuming the same composition and proportions of hot metal, but replacing 100 lb of the limestone by burnt lime, the weight of scrap charged may be reduced to 838 lb, or 830 lb of clean scrap, per ton of ingot.

Operating with oxygen lancing, the scrap, limestone, and part of the ore and lime will be quickly charged and followed by the addition of the hot metal. As soon as the initial reaction with the hot metal subsides, the fuel supply should be cut to

TABLE IX

Comparison of fuel requirements of OH furnace practice, 65% hot metal with atmospheric and oxygen-enriched air (per ton of ingot)

	Atmospheric air			Enriched air		
Heat input		10^3 Btu			10^3 Btu	
Heat in metal		711			711	
Oxidation of metal		1331			1331	
Slag formation		93			93	
Fuel	228 lb	4310		150 lb	2835	
Steam		31			21	
Regenerated air at 46·3 Btu/ft³	48500 ft³	2245		29800 ft³	1378	
		8721			6369	
Heat use			%			%
Heat in: liquid steel (1620° C)		1357	19·6		1357	26·8
slag and fume		353			353	
Decomposition of oxide and flux		498	5·7		498	7·8
Absorbed in regenerators		2245	25·8		1378	21·6
Chimney gases (750° C)	65100 ft³	1854	21·2	38550 ft³	1156	18·2
Radiation and other losses		2414	27·7		1627	25·5
		8721	100·0		6369	99·9

TABLE X

OH practice 65% hot metal, O_2 lancing

MATERIALS BALANCE			
Charged	lb	***Products***	lb
Hot metal	1500	Oxidized metal	181
,, additions			6
Clean scrap	830	Tapping loss	11
Ore Fe (237 lb)	142	Casting scrap	56
Additions	22	Ingots	2240
	2494		2494

HEAT BALANCE 10^3 Btu				
Available		***Used***		
Hot metal	711	Heat in: liquid steel	1357	22·8
Scrap		slag, etc.	353	5·9
Oxidation	1333	Decomposition of ore and fluxes	496	8·4
Slag formation	93	Absorbed in regeneration	1306	22·0
Regenerated air (28200 ft³ at 1200° C)	1306	Chimney gases (34900 ft³ at 750° C)	1047	17·6
Oxygen (650 ft³)		Radiation, etc.	1384	23·3
Fuel	2500			
	5943		5943	100·0

Heat absorbed in metal, slag, etc. 38% of heat supplied and generated

Ingot yield: 96·1% of metal charged
89·8% of total metallics charged

and on the speed of operation. The only heat available in the furnace chamber for the steelmaking operation is that heat which is not taken away by the gases.

In Table IV, which represents good modern practice with 65% hot metal fired with fuel oil, the total heat supplied as fuel, by regeneration of air and by oxidation of the bath, is given as 80 therms/ton of ingot. Of this, only 15 therms are absorbed by the metal and slag. About 41 therms are carried away by the issuing gases and of this about half is carried away by the inert N_2 gas.

As in converter practice previously discussed, the replacement of combustion air by tonnage O_2 would permit this heavy load of inert N_2 to be either completely eliminated or reduced with corresponding thermal benefit and therefore reduction of the fuel required per ton of steel.

A long series of tests carried out by the Consett Company, under Mr J. Allen, has demonstrated the value of this principle of O_2 addition to the air for combustion in accelerating the rate of production and lowering heat consumption. It has also been fully demonstrated by the published data of the National Steel Corporation of the USA.

An estimate of the benefits to be anticipated is given in Table IX which shows that with the replacement of 25% of the combustion air by tonnage O_2, giving enriched air of about 26%O_2, the fuel consumption should be reduced to about 16 gal (or 150 lb) of fuel oil per ton of ingot.

With this enrichment, the combustion air would be reduced from 48 500 ft³ to 30 000 ft³ and the volume of the waste gases would fall from 65 000 ft³ to 39 000 ft³, which means a reduction of 40% in the heat per ton of ingot carried into the slag pockets and regenerating system.

Similarly, if the temperature level of the furnace system as a whole be regarded as unchanged, the hourly losses by radiation etc. would not be changed but would show a corresponding reduction per ton of ingot made.

The work already carried out has shown that if the same hourly rates of fuel input, of combustion air, of gas velocities and temperature through the regenerator system, and radiation losses are maintained unchanged, the addition of tonnage oxygen to give about 26%O_2 in the combustion air should give a reduction in fuel consumption of about 35%. Conversely, with the same rate of fuel input per hour, the rate of production should increase by some 50% or, allowing for the time occupied in tapping and fettling, there should be an overall increase in production of about 35%.

The amount of oxygen required for this enrichment would be about 1 370 ft³ and the cost of this tonnage oxygen is offset by the saving of 8 gal of fuel oil.

No substantial capital expenditure is involved in the installation of oxygen equipment for the enrichment of combustion air and the increase in production should yield a substantial reduction in overhead and capital charges. General experience indicates that this is the most simple and direct application of tonnage oxygen to secure increased production and lower costs, particularly in existing plants.

Operating speed

The shortcomings of the OH furnace with regard to operation are too well known to require elaboration. The raw materials, pig iron, scrap, ore, and limestone are charged into the hearth, and gas or liquid fuel is burnt in the space above the charge. At a very early stage, liquid metal is added and the whole

mass of material is covered as quickly as is practical by a layer of slag. Except in the limited early stages of charging, the conditions for heat transfer and rapid reaction are extremely poor.

The heat required must pass from the flame to the metal through the slag, which is a good insulating material. Purification of the metal is effected by oxygen, either from oxide in the slag or by oxygen from the flame passing through the slag. The reaction zone is limited to a shallow area at the interface of the slag and the metal. The impurities in the metal must travel to the interface by diffusion and the oxygen at the interface must be replaced by diffusion in the slag layer.

Further, the normal oxidation by Fe_2O_3 in the metal is endothermic; the metal is cooled, and time and heat are needed to restore the required temperature conditions. These are all time factors. The reactions which take place in 12–20 min in the converter process take hours in the OH furnace, not because of any chemical basis but solely because of physical conditions.

It was demonstrated by the present author more than 30 years ago that, given the desired temperature and slag conditions, raw hot metal could be converted into steel in the OH furnace in 20 min. The only factor determining the length of time of the heat is the slow rate of heat and oxygen transmission through the slag and the relative lack of movement in the metal itself. The essentials for improvement are physical agitation of the metal and generation of heat *within the metal itself*. The advent of tonnage oxygen applied direct in the metal by lancing provides the obvious solution.

Since the oxidation of Si and P by Fe_2O_3 gives a rough thermal balance and no gas is formed, attention can be concentrated on the C/O_2 reaction. If oxygen is blown into the metal at supersonic speed the reaction is vigorous; as in the LD process, carbon is oxidized rapidly and the metal is agitated violently by both the O_2 jet and the gas evolution. The rate of elimination of carbon is limited only, and can be controlled, by the rate of oxygen-blowing. The value of this to secure low-C content of the bath has been fully demonstrated by Dr Kesterton at the Abbey works of The Steel Company of Wales.

Oxidation by O_2 lancing, as in the converter, leads to loss of yield by the loss of carbon and the metal becomes far too hot. As in the converter, *it is necessary to add a cooling agent* and in this case the agent used is iron ore which, for obvious reasons, should be of high quality. The thermal reactions involved should be examined in two stages; in the metal, and in the slag.

Metal

Injecting cold oxygen into a bath of liquid metal at, say, 1 520° C gives for each lb of carbon liberated from Fe_3C and oxidized to CO an evolution of heat in the metal of 5 160 Btu and forms 2·33 lb (31·5 ft³) of CO which will pass through the slag and burn to CO_2. This evolution of heat will obviously raise the temperature of the metal and of the slag.

On the other hand, if Fe_2O_3 is added as a 'coolant' each lb of carbon oxidized by Fe_2O_3 will require 4·44 lb of oxide and yield 3·11 lb of Fe, increasing the metal weight by 2·11 lb and, again, release 31·5 ft³ of CO. The reduction of the oxide and the heating of the metallic iron reduced will absorb roughly 11 700 Btu/lb of carbon oxidized. This is counterbalanced by the heat of oxidation of C from Fe_3C to CO of 5 160 Btu, giving a nett absorption of heat in the metal of 6 500 Btu/lb of C. This means that the metal is cooled and the reaction slowed down.

TABLE XI

OH practice, 65% hot metal, oxygen-enriched air 750 ft³ O_2, oxygen lancing 650 ft³ O_2

HEAT BALANCE, per ton of ingot 10^3 Btu

Available		*Used*	
Heat in hot metal	711	Heat in: liquid steel	1357
Oxidation	1333	slag (400 lb)	353
Slag formation	93	Decomposition of ore and fluxes	496
	2137		2206
Regenerated air (1200° C) 21700 ft³ at 46·3	1005	To regeneration chambers	1005
Fuel oil (111 lb) – 21 therms	2100	Chimney gas (750° C) 29000 ft³	870 870
Steam	15	Radiation, etc. (22·3%)	1176
	5257		5257

Heat absorbed in metal and slag: 42% of heat supplied and generated

Slag surface

In both cases the escaping CO will burn with the air in the furnace on the surface of the slag, thereby imparting heat direct to the slag. The combustion of 13·5 ft³ of CO requires 75 ft³ of air, and assuming that the air is regenerated to 1 200° C and the products of combustion (with 20% excess air) escape at 1 720° C, the nett heat evolved on the slag is about 5 300 Btu/lb of C oxidized.

Oxidation of C by Fe_2O_3 cools the metal and tends to slow down the reaction until further O_2 and heat are available. Oxidation by O_2 jet liberates heat in the metal and stimulates physical agitation and reaction. To obtain the best results, the obvious course of action is, as in the converter, to combine the two.

From the above data, each lb of carbon oxidized by Fe_2O_3 absorbs 6 500 Btu while each lb of C oxidized by oxygen jet evolves 5 600 Btu. To balance these actions, 46 lb of C should be oxidized by iron ore and 54 lb by oxygen.

The example of a typical hot-metal practice given in Table IV may be taken as an example for the application of oxygen lancing. Assuming the same composition and proportions of hot metal, but replacing 100 lb of the limestone by burnt lime, the weight of scrap charged may be reduced to 838 lb, or 830 lb of clean scrap, per ton of ingot.

Operating with oxygen lancing, the scrap, limestone, and part of the ore and lime will be quickly charged and followed by the addition of the hot metal. As soon as the initial reaction with the hot metal subsides, the fuel supply should be cut to

roughly half the previous rate, or less, and the oxygen lance applied. A vigorous reaction will be maintained and the air supply should be regulated to the amount required for the combustion of the fuel and of the CO generated by the reaction. This can be readily calculated. The whole of the burnt lime and ore required may be included in the initial charge but better results will be obtained if a substantial part of these additions is charged systematically during the progress of the heat. With the vigorous reaction which is maintained by the oxygen lance, a considerable amount of foaming of the slag may occur but this can be regulated by the application of steam jets.

A summary of the results to be anticipated is given in Table X. It will be noted that with the use of iron ore and the oxygen lance, both the amount of air passing through the regenerators and the volume of the exit gases are only a little over half of those with the normal practice. The total amount of heat per ton of metal in the system is reduced from 87 to 59·5 therms and the loss of heat in the chimney gases, assuming the same temperature, is reduced from 18·5 to 10·5 therms. The consumption of fuel is reduced from 43 to 25 therms, even after allowing for 24% of the total heat as radiation and other losses.

It therefore follows that, assuming the same rate of heat input into the furnace, the rate of production should increase in the ratio of 25 : 43, so far as the time from finished charging to tapping is concerned. Assuming no change with regard to the charging, tapping, and fettling periods, this corresponds to an increase of about 45–50% in the rate of production.

It is to be noted also that with the use of iron oxide for oxidation, the yield of metal also shows an increase by the replacement of the carbon oxidized by Fe reduced from the ore. Taken overall, with the use of about 650 ft³ of tonnage oxygen per ton of ingot, the rate of production should show an increase of some 45%, the fuel consumption should fall to about 25 therms/ton of ingot, and the yield of ingot should be of the order of 96% from hot metal and scrap charged, or 90% of the total metallics charged.

Two further points of importance are to be noted. With the use of oxygen, the total volume of gases passing through the furnace chamber, even with the rapid elimination of carbon envisaged, would be considerably less than in the normal practice. It is therefore to be anticipated that the erosion of the refractories should be reduced and roof life should be improved.

Further, with the increased rate of production, the overhead and capital charges should show the corresponding economy, in addition to the saving in fuel costs.

There are therefore two important directions in which tonnage oxygen can be applied to the OH process. Each will give a substantial saving in fuel and increase productivity, and the application of each principle is capable of wide variation. Probably the best results will be obtained by a combination of the two, in which enriched air is used during the charging and preliminary melting periods and for the finishing of the heat, while in the oxidation period practically no fuel is employed and the work of oxidation is shared between the oxygen jet and iron ore. This has been discussed by Jackson.*

In Table XI, an estimate is given of the operation of a furnace using 65% hot metal with 750 ft³ of oxygen for air enrichment and 650 ft³ for lancing. It is estimated that the fuel consumption should not exceed 21 therms and the rate of production should definitely be some 40–50% higher than with the present good practice.

As stated earlier, it is often claimed that the capital cost of a converter plant is not more than about two-thirds of the present cost of an OH plant. If OH furnaces are operated with either oxygen enrichment of the combustion air or with oxygen lancing combined with ore oxidation, or a combination of the two, the enhanced rate of production would reduce the capital expenditure per ton of output of the OH furnace to the same order as that in the converter plant. It will be accepted that these applications of oxygen present the possibility of a great improvement in OH practice. This improvement, combined with the greater flexibility of the OH furnace as regards quality of material used, variation of hot metal and scrap proportions, the wider application to the making of quality steels, and the more reliable control of the operation, will more than maintain the position of the OH process as a steelmaking operation for a wide field of steel quality. In my opinion, views recently expressed that the days of the OH furnace are past are based on an inadequate examination of the problems involved and the possibilities afforded by modern techniques, and should be subject to careful reconsideration.

ELECTRIC ARC FURNACES

The position of the electric arc furnace process must be examined from two quite distinct aspects, the manufacture of steel of special quality and of ordinary commercial types.

During the Second World War, the production of special and alloy steel showed a very substantial increase, and at the end of the war a very serious but natural decline occurred. In line with other activities, the use of these qualities of steel increased and by 1950 the wartime level of production was regained and steady progress has since been maintained. The development of the use of oxygen and other specialized techniques has been fully described in recent papers and is not within the scope of this review.

The development of arc furnaces with a capacity of 100 tons or more with the natural improvements in productivity and operating costs inevitably raises the question of the competitive strength of the arc furnace as a manufacturer of commercial qualities of steel. This question is sharpened by the fact that the new oxygen-blown converters will not absorb their proportionate arising of scrap and it may become necessary to determine how this excess scrap shall be utilized. The development of the single slag operation of the arc furnace furnishes a possible new approach.

* A. Jackson: *JISI*, 1958, vol. 190, Sept., pp. 1–29; 1959, vol. 191, Apr., pp. 337–342.

It will be accepted that the key factors in the comparative economics are the capital cost and the cost of the thermal energy consumed in the electric furnace, as compared with the OH furnace. There can be no question that the capital cost of the large electric furnace is considerably lower and that the general melting costs should be lower per ton of production than those of the OH furnace. On the other hand, the cost of electric power and electrodes, even with the low consumption figures of the single slag operation, is considerably higher than the fuel costs in the OH furnace.

The position today is marginal and is dependent upon two considerations; the price at which scrap is made available within the works economy and how far the price of that scrap should be debited with the capital charges which have been incurred in making the scrap. Looking to the future, it is certain that the manufacturing costs of OH steel will be reduced by the application of the new oxygen techniques. It is equally certain that the power requirements of the arc furnace can be reduced by oxygen lancing during the melting period and this, coupled with the anticipated reduction in power costs by the use of nuclear energy, should lead to the narrowing of the margin between the two processes. Depending upon the speed with which these improvements are effected, a wider adoption of the arc furnace for melting available scrap supplies may be looked for.

CONCLUSION

The outstanding feature of development in steelmaking during the post-war period is the application of tonnage oxygen.

The use of oxygen with steam in the traditional Thomas process reduces or eliminates the deleterious N_2 and reduces the gap between commercial Thomas and OH quality.

The development of the new LD process has made it practical to extend the pneumatic process to the manufacture of steel from pig iron of intermediate P content. This steel is of very low N_2 content and of OH quality.

In both cases the competitive strength of the pneumatic processes has improved. With the growth of the steelmaking industry in the UK, a substantial tonnage of the additional imported ore required will be of low P content. Pig iron made from this ore, alone or mixed with other ore, will be in the range which is eminently suitable for the LD process, and there can be no doubt that this process will be adopted in the UK particularly for the manufacture of deep-drawing and strip mill qualities.

The inherent handicap of the OH furnace in respect of high fuel consumption, low thermal efficiency, and low productivity, can be revolutionized by the application of oxygen for the pre-refining of the iron, the enrichment of the combustion air, and/or the direct refining of the metal in the furnace. This application will reduce operating costs, particularly of fuel consumption, and should increase productivity by 40–50% without incurring substantial expenditure.

If these modern techniques are adopted, the OH process should be fully capable of maintaining its competitive position as regards both flexibility and quality control. In addition, the growing demand for increased production and the elimination of obsolescent plant can be met at relatively low capital cost.

APPENDIX A

Acid Bessemer practice, per ton of ingot (mild steel)

Hot metal charged: 1·118 tons or 2505 lb
Scrap added: 45 lb

A Oxidation

	C	Si	Mn	S	P	Fe	Total
Hot metal, %	4·3	2·0	0·3	0·04	0·04		
Blown metal, %	0·1	...	0·1	0·04	0·04		
Oxidized, lb	105·4	50·1	5·2	...	...	107·8	268·5
O_2 required, lb	140·5	57·3	1·5	...	...	30·8	230·1

B Metal balance

		lb	lb
Charged	hot metal	2505	
	scrap	45	2550
Less	metal oxidized	268·5	
	ejections (0·5%)	12·5	281
Blown metal			2269
Additions, 20 lb less oxidation			16
Liquid steel in ladle			2285
Casting scrap			45
Ingot weight			2240
Yield	ingots from hot metal and scrap charged		87·8%
	ingots from metallics charged		87·1%

C Slag

			lb	%
SiO_2	from metal	107·4		
,,	,, vessel	22·0	129·4	53·3
MnO	,, metal		6·7	2·7
FeO	,, ,,		106·5	44·0
			242·6	100·0

D Air blown and waste gases

O_2 required: 230·1 lb	2720 ft³		
Air blown at 21% O_2	12950 ft³	%	
Nitrogen in air blown	10230 ft³	75·5	
CO formed	3320 ft³	24·5	
Waste gases	13550 ft³		

E Heat balance

Available

			Btu/lb	10^3 Btu	
Heat in liquid metal (1350° C)	2505 lb	at	474	1187·3	
Oxidation:					
C	105·4	at	4350	458·5	
Si	50·1	at	12176	610·0	
Mn	5·2	at	3160	16·4	
Fe	107·8	at	2067	222·8	
Fe_3C	105·4	at	810	85·4	
Formation of silicate	106·5	at	147	15·6	
				2596·0	

Used

Heat in:					
steel (1620° C)	2270	at	562	1275·7	
slag	243	at	820	199·3	
fume	32	at	820	26·2	%
				1501·2	57·8
gases (1720° C)					
N_2	10230	at	67·0	685·4	
CO	3320	at	67·7	224·7	
				910·1	35·1
Radiation and not accounted for				184·7	7·1
				2596·0	100·0

APPENDIX B

Basic Bessemer practice, per ton of ingot (mild steel)

Hot metal charged: 2500 lb
Scrap added (2%): 45 lb 2545 lb

A Oxidation

	C	Si	Mn	S	P	Fe
Hot metal 2500 lb, %	3·6	0·3	0·8	0·05	1·80	
Blown metal, %	1·05	...	0·1	0·03	0·04	
Oxidized, lb	88·9	7·5	17·8	0·55	44·1	104·1
O_2 required, lb	118·5	8·6	5·2	...	56·9	29·7
Product, CO	207·4	16·1	23·0	0·55	101·0	133·8

B Metal balance

			lb
Charged	hot metal	2500	
	scrap	45	2545
Less	metal oxidized	263	
	ejections (0·5%)	13	276
Blown metal			2269
Additions 20 lb less loss			16
Liquid steel			2285
Casting scrap (2%)			45
Ingots			2240

Yield ingots from hot metal plus scrap 88·0%
ingots from total metallics 87·3%

C Slag

	lb	lb	%
SiO_2 from metal	16·1		
„ „ fluxes	12·5	28·6	5·4
Al_2O_3		4·2	0·8
P_2O_5		101·0	19·2
CaO		245·3	46·7
MgO		17·1	3·2
MnO		23·0	4·4
CaS		4·5	0·9
FeO		101·7	19·4
		525·4	100·0

D Air blown and waste gases

Oxygen required: 218·9 lb	2588 ft³	
Air blown	12324 ft³	
Gases:		%
N_2	9736 ft³	77·7
CO	2800 ft³	22·3
	12536 ft³	100·0

E Heat balance

Available

Heat in hot metal:			Btu/lb	10³ Btu	
	2500 lb	at	474	1185·0	
Oxidation:					
C	88·9	at	4350	386·7	
Si	7·5	at	12176	91·3	
Mn	17·8	at	3160	56·2	
P	44·1	at	10843	478·2	
Fe	104·1	at	2067	215·2	
Fe_3C	88·9	at	810	72·0	
Slag formation:					
2 CaO, SiO_2	28·6	at	866	24·7	
3 CaO, P_2O_5	101·0	at	2020	204·0	
				2713·3	

Used

Heat in:					
Liquid steel	2269	at	562	1275·2	
Slag	525	at	820	430·5	
Fume	32	at	820	26·2	%
				1731·9	63·8
Gases at 1720° C					
N_2	9736	at	67·0	652·3	
CO	2800	at	67·7	189·6	
				841·9	31·0
Radiation and not accounted for				139·5	5·1
				2713·3	99·9

APPENDIX C

OH practice cold charging, (30% pig iron)

Charge per ton of ingot:

Scrap:	1707 lb less 1% sand	1690 lb	
Pig iron:	735 lb	735 lb	2425 lb

A Oxidation

	C	Si	Mn	S	P	Fe	Total
Pig iron 735 lb, %	3·8	1·0	0·8	0·06	1·4		
Scrap 1690 lb, %	0·15	0·03	0·55	0·05	0·05		
Tap, %	0·1	...	0·15	0·04	0·04		
Oxidized, lb	28·17	7·85	11·75	0·37	10·23	75·6	133·97
O_2 required, lb	75·12	8·97	3·42	...	13·20	21·6	122·3
Product (CO_2)	103·3	16·8	15·2	0·37	23·43	97·2	

B Metal balance

		lb
Metal charged		2425
Less oxidized		134
Bath at tapping		2291
Additions (nett)		16
		2307
Less tapping loss	11 lb	
„ casting scrap	56 lb	67 = 2240 lb
Ingot yield	from metal charged	92·3%
	from total metallics	91·6%

C Slag formed

SiO_2	Al_2O_3	P_2O_5	CaO	MgO	MnO	CaS	FeO	Total
41·7	5·7	23·4	137·4	21·3	15·2	2·1	69·7	316·5 lb
13·1	1·8	7·4	43·4	6·7	4·8	0·7	22·0	99·9%

Fumes: FeO 27·5 lb CaO 4·0 lb

D Heat balance

Available

Generated in bath:			Btu/lb	10³ Btu
C	28·17	at	14550	409·9
Si	7·85	at	12176	95·6
Mn	11·75	at	3160	37·1
P	10·23	at	10843	110·9
Fe_3C	28·17	at	810	22·8
Fe(S) 54·2, Fe(F) 21·4		at	2067	156·3
Total heat from meta				832·6
Slag formation:				
2 CaO, SiO_2	41·7	at	809	33·7
3 CaO, P_2O_5	23·43	at	2020	47·3
				913·6
Heat in regeneration (1200° C)				
air	42315	at	46	1946·0
gas	27657	at	46	1273·0
Producer gas				4500·0
				8632·6

Used

Heat in:				
Metal	2291	at	562	1287·5
Slag	317	at	820	259·5
Fume	30	at	820	24·6
				1571·6
Decomposition of $CaCO_3$	59	at	1373	81·0
Absorbed in regeneration				3219·0
Chimney gas at 750° C	71800			2154·0
Radiation loss and not accounted for (18·6%)				1607·0
				8632·6

APPENDIX D

OH practice 65% hot metal

		lb		lb
Charge:	Hot metal	1500	Ore (60%)	200
	Scrap (1% SiO_2)	858	Limestone	160
		2358	Lime	75
	Additions	20	Dolomite	60

A Oxidation

	C	Si	Mn	S	P	Fe
Hot metal 1500 lb, %	4·0	0·5	0·8	0·05	0·65	
Scrap 850 lb, %	0·05	...	0·35	0·05	0·05	
At tap, %	0·05	...	0·15	0·025	0·015	
Oxidized, lb	59·3	7·5	11·6	0·6	9·8	90·2
O_2 required, lb	158·0	8·6	3·3	...	12·7	25·8
Product (CO_2)	217·3	16·1	14·9	0·6	22·5	116·0

Total O_2 required 208·4 lb
Less O_2 in Fe_2O_3 51·4 lb
O_2 from air 157·0 lb = 1856 ft³
Air required (oxidation) 8840 ft³

B Metal balance

	lb		lb
Hot metal	1500	Oxidized metal	179
Scrap	850	,, additions	4
Oxides Fe (200 lb)	120	Casting scrap	67
Additions	20	Ingots	2240
	2490		2490

C Slag formed

	SiO_2	Al_2O_3	P_2O_5	CaO	MgO	MnO	CaS	FeO
Weight (400 lb)	55·0	9·5	22·5	185·9	22·8	14·9	2·7	86·5
Wt-%	13·7	2·4	5·6	46·5	5·7	3·8	0·7	21·6

Fe in slag: 16·7% FeO in fume, etc: 29·5 lb

D Heat balance

Available			Btu/lb	10^3 Btu	
Hot metal	1500	at	474	711·0	
Oxidation:					
C	59·3	at	14550	862·8	
Si	7·5	at	12176	91·3	
Mn	11·5	at	3160	36·3	
P	9·8	at	10843	106·2	
Fe	90·2	at	2067	186·4	
Fe_3C	59·3	at	810	48·0	
				1331·0	
Slag formation: 2 CaO, SiO_2	55·0	at	866	47·6	
3 CaO, P_2O_5	22·5	at	2020	45·4	
				93·0	
Regenerated air (1200° C)	48500	at	46·3	2245·5	
Fuel				4310·0	
Steam				31·0	
				8721·5	
Used					
Heat in:					%
Liquid steel	2291	at	562	1357·4	15·6
Slag and fume	430	at	820	352·6	4·0
				1710·0	19·6
Decomposition:					
Fe_2O_3	120	at	3191	382·9	5·7
$CaCO_3$	150	at	768	115·2	
Absorbed in regenerators				2245·5	25·8
Chimney gases at 750° C	65100			1854·0	21·2
Radiation and not accounted for				2413·9	27·7
				8721·5	100·0

APPENDIX E

Typical tilting OH practice

Charge per ton of ingot:

Hot metal	1906 lb	Ore (65% Fe)	250 lb
Scrap	398 lb	Limestone	49 lb
	2304 lb	Burnt lime	22·4 lb
Finishings	22·4 lb	Dolomite	71·7 lb

A Oxidation

	C	Si	Mn	S	P	Fe
Hot metal 1906 lb, %	3·01	0·22	0·47	0·05	0·91	
Scrap 398 lb, %	0·20	...	0·55	0·06	0·06	
Tapping, %	0·20	...	0·07	0·04	0·046	
Oxidized, lb	53·6	4·2	9·6	0·24	16·5	59·0
O_2 required, lb	142·9	4·8	2·8	...	21·3	16·9
Product, (CO_2)	196·5	9·0	12·4	0·24	37·8	75·9

Total O_2 required 188·7 lb
Less O_2 from Fe_2O_3 69·6 lb
O_2 from air 119·1 lb 1408 ft³
Air required 7000 ft³

B Metal balance

	lb		lb
Hot metal	1906	Oxidized metal	143
Scrap	398	,, additions	6
Oxides (250 lb at 65%)	162	Castings scrap, etc.	99
Additions	22	Ingots	2240
	2488		2488

C Slag formed

SiO_2	Al_2O_3	P_2O_5	CaO	MgO	MnO	CaS	FeO	
15·4	1·7	37·8	86·8	27·2	12·4	1·0	75·9	lb
6·0	0·6	14·65	33·65	10·5	4·8	0·3	29·5	%

Slag weight: 260 lb (Fe: 22·9%)

D Heat balance

Available			Btu/lb	10^3 Btu	
Heat in metal	1906	at	474	903·4	
Oxidation:					
C	53·6	at	14550	779·9	
Fe_3C	53·6	at	810	43·4	
Si	4·2	at	12176	51·1	
Mn	9·6	at	3160	30·3	
P	16·5	at	10843	178·9	
Fe	59·0	at	2067	122·0	
				1205·6	
Slag formation:					
2 CaO, SiO_2	15·4	at	866	13·3	
3 CaO, P_2O_5	37·8	at	2020	76·4	
				89·5	
Regeneration (1200° C)	49060	at	46·3	2271·7	
Fuel oil				4600·0	
				9070·2	
Used					
Heat in:					
Liquid steel	2342	at	562	1316·2	14·5
Slag	260	at	820	219·8	2·4
Decomposition of:					
$CaCO_3$	46	at	768	35·3	6·1
Fe_2O_3	1625	at	3191	518·5	
Regeneration				2271·5	25·1
Chimney gas (850° C)	63375	at	35	2218·0	24·4
Radiation and not accounted for				2490·9	27·4
				9070·2	99·9

APPENDIX F

Basic Bessemer practice, low phosphorus iron, per ton of ingot (mild steel)

Hot metal charged:	2443 lb	1·091 tons
Scrap charged:	45 lb	0·020 tons
	2488 lb	1·111 tons

A Oxidation

	C	Si	Mn	S	P	Fe	Total
Hot metal, %	4·2	0·6	1·0	0·06	0·30		
Blown metal, %	0·05	...	0·10	0·04	0·04		
Oxidized, lb	101·4	14·6	22·2	0·56	6·4	61·4	206·6
O_2 required, lb	135·2	16·7	6·4	...	8·2	17·6	184·1
Product (CO)	236·6	31·3	28·6	0·56	14·6	79·0	

B Metal balance

		lb
Charged hot metal	2443	
scrap	45	2488
Less oxidized	206·6	
ejections (0·5%)	12·4	219
Blown metal		2269
Additions less oxidation		16
Liquid steel in ladle		2285
Casting scrap (2%)		45
Ingots		2240

Yield: ingots from hot metal and scrap 90·0%
ingots from total metal charged 89·3%

C Slag formed

	lb	%
SiO_2 metal	31·3	
,, fluxes	6·2	
	37·5	14·0
Al_2O_3	2·0	0·8
P_2O_5	14·6	5·5
CaO	116·4	43·6
MgO	17·1	6·4
MnO	28·6	10·7
CaS	2·8	1·1
FeO	47·6	17·9
	266·6	100·0

D Air blown and gases

Oxygen required	ft³	%
184·1 lb	2176	
Air required	10360	
N_2 from blast air	8184	71·9
CO from metal	3197	28·1
Waste gases	11381	100·0

E Heat balance

Available			Btu/lb	10^3 Btu
Hot metal	2443	at	474	1158·0
Oxidation:				
C	101·4	at	4350	441·1
Si	14·6	at	12176	177·7
Mn	22·2	at	3160	70·1
P	6·4	at	10843	69·4
Fe	61·4	at	2067	126·9
Fe_3C	101·4	at	810	82·1
Slag formation:				
2 CaO, SiO_2	37·5	at	866	32·5
3 CaO, P_2O_5	14·6	at	2020	29·5
				2187·3
Deficit 10%				222·5
				2409·8

Used				
Heat in:				
Liquid steel	2269	at	562	1275·2
Slag	267	at	820	218·9
Fume	31·4	at	820	25·7
				1519·8
Gases (1720° C)				
N_2	8184	at	67·0	548·3
CO	3197	at	67·7	216·4
				764·7
Radiation, etc. at 5·2%				125·3
				2409·8

APPENDIX G

LD process, low phosphorus iron (per ton of mild steel)

Charge: Hot metal 2360 lb
Scrap 45 lb
Iron ore (56% Fe) 157 lb

A Oxidation

	C	Si	Mn	S	P	Fe
Hot metal, %	4·2	0·6	1·2	0·06	0·30	
Blown metal, %	0·05	...	0·10	0·04	0·04	
Oxidized, lb	98·0	14·2	26·1	0·5	6·2	67·0
O_2 required, lb	130·7	16·2	7·6	...	8·0	19·1
Product, (CO)	228·7	30·4	33·7	0·5	14·2	86·1

Total O_2 required 181·6 lb
Less O_2 in Fe_2O_3 37·7 lb
O_2 required to be blown 143·9 lb = 1700 ft³
CO formed 228·7 lb = 3090 ft³

B Metal balance

	lb		lb
Hot metal	2360	Oxidized metal	213
Scrap	45	„ additions	4
Fe from ore	89	Ejections	12
Additions	20	Casting scrap	45
		Ingots	2240
	2514		2514

Yield of ingots: 93·1% of metal charged
89·1% of total metallics

C Slag formed

		lb	%
SiO_2 from metal	30·4		
from ore	19·0		
from lime	6·1	55·5	18·5
Al_2O_3		8·4	2·8
P_2O_5		14·2	4·7
CaO		127·8	42·6
MgO		2·5	0·8
MnO		33·7	11·2
CaS		3·6	1·2
FeO		54·0	18·0
		299·7	99·8

Lime added: 136 lb Fe in slag: 14·0%

D Heat balance

Available			Btu/lb	10^3 Btu
Heat in hot metal	2360	at	474	1118·6
Oxidation:				
C	98·0	at	4350	426·3
Si	14·2	at	12176	172·9
Mn	26·1	at	3160	82·5
P	6·2	at	10843	67·2
Fe_3C	98·0	at	810	79·4
Fe	67·0	at	2167	145·2
Slag formation:				
2 CaO, SiO_2	55·0	at	866	47·6
2 CaO, P_2O_5	14·2	at	2020	28·7
				2168·4

Used					
Heat in:					%
Liquid steel	2269	at	562	1275·2	58·8
Slag and fume	340	at	820	278·8	12·9
				1554·0	71·7
Gases:					
CO	3090	at	67·7	209·9	9·7
N_2	10	at	67·0		
Reduction of Fe_2O_3	88	at	3190	283·9	13·0
Radiation and losses				120·6	5·6
				2168·4	100·0

Useful heat added to ingot, metal, and fume 41·5% of heat generated
Heat usefully employed 68·2% of heat generated

APPENDIX H

OH process 65% hot metal, O_2 lancing

Charge:

Hot metal	1500 lb	Ore (60% Fe)	237 lb
Scrap (1% SiO_2)	838 lb	Limestone	60 lb
	2338 lb	Burnt lime	110 lb
Additions	22 lb	Dolomite	60 lb
		O_2 lance	650 ft³/ton

A Oxidation

		C	Si	Mn	S	P	Fe
Hot metal	1500 lb, %	4·0	0·5	0·8	0·05	0·65	
Scrap	830 lb, %	0·15	...	0·55	0·06	0·06	
At tap	2391 lb, %	0·1	...	0·10	0·025	0·015	
Oxidized, lb		58·8	7·5	14·1	0·65	9·9	90·2
O_2 required, lb		78·4	8·6	4·1	...	12·8	25·8

Total O_2 required in bath 129·7 lb
O_2 released from Fe_2O_3 61·0 lb
O_2 required from air 68·7 lb
Assume 80% from O_2 lance 55·0 lb or 650 ft³
Oxygen from air for bath 13·7 lb
for CO 78·4 lb
92·1 lb = 1089 ft³
Air for oxidation 5186 ft³
Air for combustion 25 therms 23000 ft³
Total air through regenerators 28186 ft³

B Metal balance

	lb		lb
Hot metal	1500	Oxidized metal	181
Clean scrap	830	„ additions	6
Fe from ore	142	Casting scrap and loss	67
Additions	22	Ingots	2240
	2494		2494

Yield of ingots: 96·1% of metal charged
89·8% of total metallics

C Slag

400 lb

D Gases

Waste gases from combustion of fuel 24350
from excess air 4600
from combustion of CO 5950 34900 ft³

E Heat balance

Available			Btu/lb	10^3 Btu
Hot metal	1500	at	474	711·0
Oxidation:				
C	58·8	at	14550	855·5
Si	7·5	at	12176	91·3
Mn	14·1	at	3160	44·5
P	9·9	at	10843	107·3
Fe_3C	58·8	at	810	47·6
Fe	90·2	at	2067	186·4
				1332·6

Slag formation:				
2 CaO, SiO_2	55·0	at	866	47·6
3 CaO, P_2O_5	22·7	at	2020	45·8
				93·4
Regenerated air (1200° C)	28200	at	46·3	1306·0
Fuel and steam				2500·0
				5943·0
Used				
Liquid steel	2291	at	562	1357·4
Slag and fume	430	at	820	352·6
				1710·0
Decomposition:				
Fe_2O_3	142	at	3191	453·0
$CaCO_3$	56	at	768	43·0
				496·0
Absorbed in regeneration				1306·0
Chimney gases (750° C)	34900			1047·0
Radiation and other losses (23·3%)				1384·0
				5943·0

APPENDIX I

OH practice, 65% hot metal O_2 enrichment and lancing

Oxygen for enriched air 750 ft³/ton
Oxygen for lancing 650 ft³/ton

A Oxidation

Total O_2 required in bath		129·7 lb		
Less O_2 from Fe_2O_3	61·0			
„ from lance	55·0	116·0 lb		
O_2 required from air for bath		13·7 lb		
O_2 required „ for CO → CO_2		78·4 lb		
Total O_2 from air		92·1 lb	or	1089 ft³
Air required for oxidation				5186 ft³
Waste gas from oxidation			N_2	4097 ft³
			CO_2	1854 ft³
				5950 ft³

B Combustion

Air required 21 therms at 920 ft³/therm	19320 ft³
Less air 750 ft³ O_2 enrichment	3510 ft³
Nett air for combustion	15750 ft³
Total air required, combustion and oxidation	20950 ft³
Air through regenerators air	20950 ft³
O_2	750 ft³
	21700 ft³ (about 23·7% O_2)

C Waste gases

	CO_2	H_2O	N_2	Air
Fuel 111 lb	2892	2325	15251	
Steam	...	1470	...	
Excess air (20%)	...	...	...	3864
Oxidation	1854	...	4097	...
	4746	3795	19348	3864
Less N_2 = 750 O_2	...	...	2821	
	4746	3795	16527	3864 = 28932 ft³

D Heat balance

Available	10^3 Btu	*Used*	10^3 Btu
Heat in hot metal	711	Heat in Liquid steel	1357
Oxidation (as appendix H)	1333	Slag and fume	353
Slag formation	93	Decomposition of ore and fluxes	496
	2137		2206
Regenerated 'air' 21700 ft³ at 1200° C	1005	Absorbed in regeneration	1005
Fuel 21 therms	2100	Chimney gas (750° C) 29000 ft³	870
Steam	15	Radiation and losses (22·3%)	1176
	5257		5257

Heat absorbed in metal and slag = 42% of heat supplied and generated

The effect of copper on martensite breakdown and hardenability of a 0·3% C steel

By A. S. Kenneford, M.Sc., F.I.M., Assoc.I.Mech.E.

A RECENT US PAPER[1] on the effect of copper on the hardenability of a medium-carbon steel concluded that Cu had only a slight effect on hardenability and that it was of little interest as a possible alloying element in very high-tensile steels.

Manuscript received on 23rd April, 1958.
The author is with the Ministry of Supply, ARDE, Fort Halstead, Kent.

TABLE I
Composition of steels used

Composition, %					
C	Si	Mn	S	P	Cu
0·30	0·06	0·39	0·030	0·032	Nil
0·30	0·01	0·39	0·020	0·018	0·54
0·30	0·06	0·39	0·017	0·020	1·04
0·31	0·16	0·40	0·017	0·018	1·59
0·31	0·16	0·27	0·018	0·025	1·96
0·31	0·10	0·37	0·017	0·022	2·53
0·30	0·11	0·42	0·018	0·021	2·97

SYNOPSIS

In connection with work on high-tensile steels, the effect of copper on martensite breakdown and hardenability of a 0·3%C steel has been examined. It has been found that copper additions raise the end temperature of martensite breakdown and that the effect on the hardenability of the steels used was about the same as that of nickel. 1574

As these conclusions are directly contrary to those arrived at from a somewhat similar investigation carried out in ARDE, it was though that the latter should be placed on record.

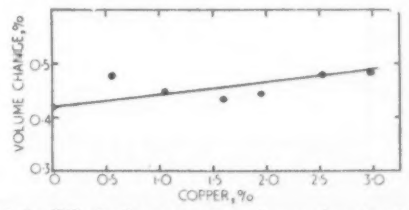

Fig. 1—Effect of copper on volume change at Ac_3

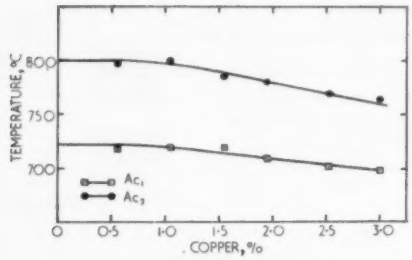

Fig. 2—Effect of copper on temperature of Ac_1 and Ac_3

MATERIALS USED

The materials used in this investigation were in the form of bar hot-rolled from small hf melts and had the compositions shown in Table I.

Residual Ni was about 0·1% and Cr 0·05% in all the steels, which were Al treated and had a grain size of ASTM 7.

DILATOMETRIC ANALYSIS

Measurements of the Ac_1 and Ac_3 temperatures were made using a Leitz dilatometer with heating and cooling rates of 5° C/min and a maximum temperature of 950° C. The effect of Cu on the volume change during the $\alpha \rightarrow \gamma$ transformation was also estimated from the dilatometric curves. Whilst these volume changes are not absolute values, they should be comparable one with another and give a general picture of the effect of Cu additions.

The effect on martensite breakdown was also determined dilatometrically under continuous heating conditions of 2·5° C/min up to a maximum temperature of 550° C in a manner previously described.[2]

The results of these experiments are presented graphically in Figs. 1–4 and show that:

(i) Copper slightly increased the volume change at the $\alpha \rightarrow \gamma$ transformation

(ii) Copper lowered the Ac_3 transformation temperature by roughly 10° C for every 1% addition up to 3%

(iii) The volume change (contraction) on tempering quenched specimens was increased by increasing Cu additions

(iv) The temperature of the end of the third stage of martensite breakdown (formation of ferrite and cementite) was raised by additions of Cu.

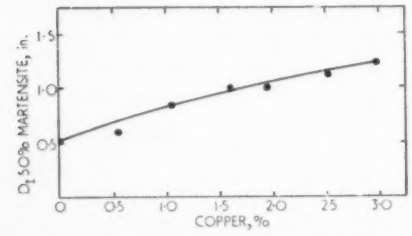

Fig. 3—Effect of copper on volume change after tempering at 550° C

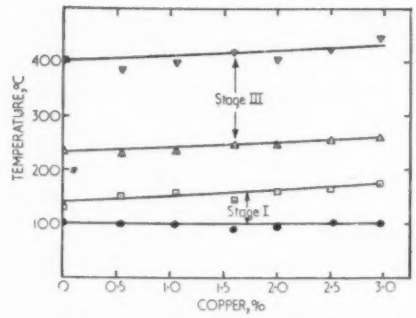

Fig. 4—Effect of copper on temperature of martensite breakdown

EFFECT ON HARDENABILITY

End-quench hardenability tests were made under standard conditions (SAE Handbook, 1947) on specimens soaked for 1 h in a protective atmosphere at a temperature 50° C above their respective Ac_3 points.

Figure 5 illustrates the results obtained and shows that Cu additions increased the overall hardness of the test bars as well as their hardenability. The curve for the plain carbon steel has been omitted as it was very close to that for the steel containing 0·5% Cu. The curves obtained with the steels containing 2·5% and 3%Cu are interesting in that they show secondary hardening caused, presumably, by the rate of cooling being such as to allow Cu precipitation in a form which caused the hardness to rise in that particular region of the test bar. In the case of the test specimens containing 2%Cu, or less, the rate of cooling over the whole length of the bar was evidently sufficient to retain all the added Cu in solution.

From the results of these tests an ideal critical diameter D_I was determined for each steel from published correlations.[3,4] These values for 50% martensite are plotted in Fig. 6, whilst the multiplying factors for hardenability calculations are given in Fig. 7. These multiplying factors obtained during the present work are not greatly different from those quoted by Kramer *et al.*,[5] and are in general similar to those for Ni.[6]

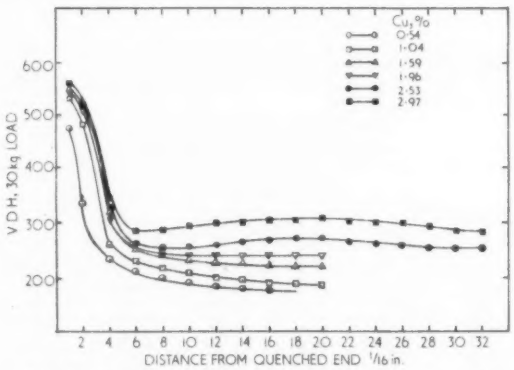

Fig. 5—End-quench hardenability of copper steels

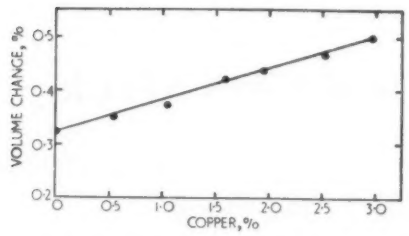

Fig. 6—Effect of copper on D_I for 50% martensite

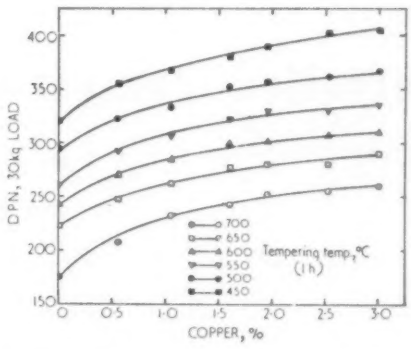

Fig. 8—Effect of copper on hardness after quenching and tempering 1 h

EFFECT ON HARDNESS AFTER TEMPERING

In order to determine the effect of tempering on the hardness of quenched Cu-bearing steels, a series of specimens $\frac{3}{4}$ in. dia × $\frac{3}{8}$ in. thick were water quenched from the appropriate temperatures and tempered for 1 h at temperatures up to 700° C. As the results were somewhat voluminous, they are not reproduced in full, but are summarized for some of the higher tempering temperatures in Fig. 8. The values obtained showed that, under the experimental conditions used, increasing Cu content up to 3% caused a progressive increase in hardness for all tempering temperatures above about 350° C. After tempering at 700° C for 1 h, the 3%Cu steel, for example, was 60 DPN harder than the plain carbon steel similarly tempered.

An interesting point noticed during tempering experiments on these quenched Cu steels was that it was impossible with tempering times up to 6 h to obtain any actual rise in hardness in the range of tempering temperature over which Cu was precipitated, though a decrease in the slope of the hardness/tempering-temperature curve was apparent. On the other hand, tempering of the normalized material resulted in the well known increase in hardness obtainable after, say, 4 h at 500° C.

The reason for this difference in behaviour between the quenched and tempered, and normalized and tempered steels is not obvious and is being investigated. 0·3%C steel has been investigated. The effects were generally found to be similar to those of Ni as far as hardenability was concerned, but unlike this latter element, Cu raised the end temperature of martensite breakdown and thus delayed softening during tempering in a manner similar to that of Si. The Cu additions were, however, not so effective in this respect as those of Si.[2] Copper also delayed softening in quenched and tempered steels by secondary hardening, and for tempering temperatures over 350° C (time 1 h) steels with 2%Cu or more resulted in an increase of some 60 DPN over the hardness of a Cu-free steel similarly treated.

It was concluded from these results that Cu should be useful as an alloying element in very high-tensile steels, and work in hand in ARDE has gone far to confirm this conclusion.

CONCLUSIONS

The effect of up to 3%Cu on the martensite breakdown temperature and hardenability of a

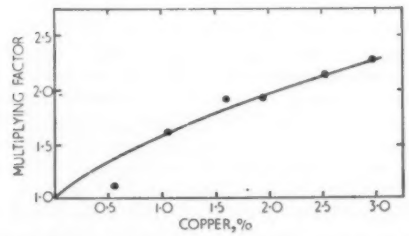

Fig. 7—Multiplying factors for copper in a 0·3%C steel

ACKNOWLEDGMENTS

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Controlled-transformation stainless steels

By K. J. Irvine, B.Sc., Ph.D., D. T. Llewellyn, B.Sc., and F. B. Pickering, A.Met., A.I.M.

SYNOPSIS

Where high-strength stainless steels are required, the choice has been limited to cold-worked austenitic or ferritic types, or to transformable compositions. These steels present difficulty where fabrication is necessary, and for many aircraft or missile applications suitable steels were not available. A considerable amount of work has now been carried out on controlled-transformation steels which have, as an essential feature, a martensite transformation range which occurs below room temperature. Consequently, the steel is austenitic initially and can be fabricated quite readily. A subsequent refrigeration treatment at — 70° C will produce transformation to martensite, or alternatively, a heat-treatment at 700° C will precipitate alloy carbides and raise the martensite transformation range above RT so that transformation to martensite will occur. Using either treatment the fabricated steel can be considerably strengthened to produce a tensile strength of 70 tons/in² with a high proof stress/tensile strength ratio.

From the work which has been carried out it has been shown that satisfactory controlled-transformation steels can be developed from a 17%Cr–4%Ni base composition using single additions of Mn, Mo, Cu, Si, W or V. When complete transformation to martensite is obtained the mechanical properties are very satisfactory, but any untransformed austenite seriously reduces the PS/TS ratio. The main limitation of these simple controlled-transformation steels is that accurate control of the martensite transformation is difficult because of the inevitable range of analysis in commercial steelmaking. It has been found, however, that the solution-treatment temperature can provide a sensitive control on analysis (and hence transformation temperature) because this controls the amount of carbide which is taken into solution.

It has also been found that certain alloying elements promote a strong age-hardening reaction and Cu, Al, Co, Ti, and Nb have been studied. The ageing reaction leads to an increase in the strength of the martensite structure and it is found that the final properties are less dependent on a complete transformation to martensite. Controlled-transformation ageing steels are therefore more suitable for commercial development and several suitable compositions are given. 1567

IN RECENT YEARS there have been some interesting developments in high-strength stainless steel. The demand for these steels arose in the aircraft industry where the continuing development of high-speed aircraft and missiles had raised the operating temperature above the maximum allowable temperature for high-strength Al or Mg alloys, i.e. above 200° C. Interest then turned to high-strength Ti alloys and stainless steel, and it was shown that martensitic stainless steels had very satisfactory properties over a wide temperature range. Consequently, there was a demand for a stainless steel which could be fabricated and welded and would retain high tensile strength to moderate operating temperatures, i.e. up to 400° C. To have a strength/weight ratio equivalent to these Al alloys the steel would need a minimum tensile strength of 70 tons/in² and none of the standard grades of stainless steel could meet these property requirements. The main types available could be classified as follows:

(i) Non-transformable: (*a*) ferritic, e.g. 17%Cr, (*b*) austenitic, e.g. 18–8. These types of steel can be strengthened by cold work.

(ii) Transformable: martensitic 12%Cr steels.

These steels were quite satisfactory for many applications and the required tensile strength was easily obtainable. The major disadvantage lay in the difficulty of fabrication. The non-transformable types possess little ductility in the cold-worked condition, and consequently it is not possible to fabricate after the material has been raised to 70 tons/in² by cold work. On the other hand, it is not possible to ensure uniform working on a complicated structure to strengthen the steel by the deformation during fabrication. The other disadvantage of this type of steel is that the enhanced properties due to cold work are lost at elevated temperatures. The transformable steels are easily fabricated in the fully softened condition, but heat-treatment of a complicated fabricated shape is difficult. The solution temperature necessary, which is of the order of 1 050° C, can cause serious scaling, and it is difficult to prevent distortion of the structure whilst at the solution-treatment temperature. Additionally, transformation stresses can cause further distortion of the structure. On the other hand, in the heat-treated condition the steel does not have sufficient ductility to allow fabrication.

A suitable steel would be one which was soft and ductile when in sheet form, and hence could be easily fabricated, but which could be hardened after fabrication by a simple type of heat-treatment which would not involve high temperatures. The obvious solution would appear to be an age-hardening mechanism and there is, of course, considerable experience in the aircraft industry with age-hardened Al alloys. These can be fabricated easily and the subsequent heat-treatment produces no scaling and little distortion. Translated to stainless steel, this would mean an austenitic structure in the sheet form which would fabricate easily and then precipitation hardening to raise the tensile strength. A considerable amount of

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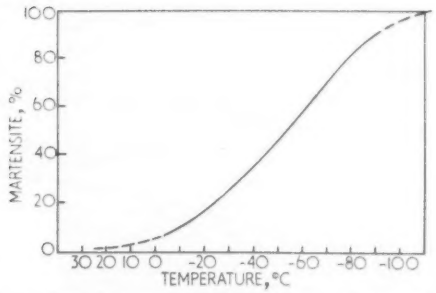


Fig. 1—Typical martensite transformation curve

work has been carried out on age-hardening austenitic alloys,[1] but, although quite high mechanical properties can be obtained, these steels do not appear to be very suitable for aircraft applications. One major objection is that they generally need high solution temperatures. In the course of work on precipitation-hardening stainless steels[2] it was considered that precipitation occurred more readily from a martensitic than an austenitic matrix, and when a precipitation-hardening mechanism was added to the strength of a martensitic structure very satisfactory tensile properties could be obtained. To meet the requirements of the aircraft industry, therefore, an austenitic steel was needed which could be easily fabricated and then transformed to a martensitic structure by a relatively simple heat-treatment. A final ageing treatment could then produce the maximum mechanical properties.

It is along these lines that most developments have taken place in the last few years. In the course of this work some confusion has arisen as to the mechanism by which the improved properties were obtained, and no clear distinction has been drawn between properties obtained by transformation and properties obtained by ageing. This work has, however, led to the production of steels with extremely attractive mechanical properties, and there seems little reason to doubt that the field of application will be widened considerably beyond the aircraft industry so that these steels may well become standard engineering constructional materials. Consequently, it was thought that it would be useful to review the types of steel which are possible and to explain the mechanisms involved.

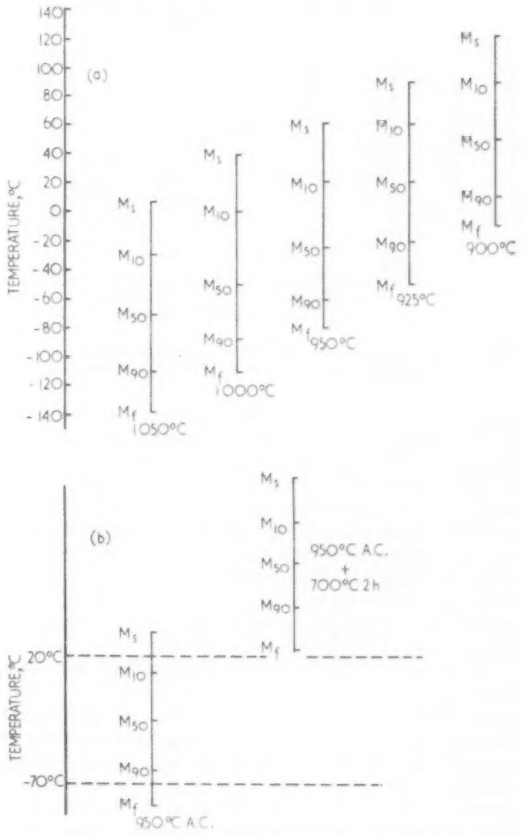
Taking into account recent developments, the types of high-strength stainless steel which can be produced are as follows:

(i) Non-transformable: (*a*) ferritic, (*b*) austenitic, and (*c*) austenitic with age hardening. Of these, (*a*) and (*b*) can be strengthened only by cold work; (*c*) can be hardened by heat-treatment and this is also aided by cold work.

(ii) Transformable: (*a*) a martensitic steel possibly with enhanced tempering resistance, (*b*) a martensitic steel with the addition of an age-hardening mechanism.

(iii) Controlled-transformation stainless steels: (*a*) simple compositions where the strength is due entirely to the transformation, but perhaps with added tempering resistance, (*b*) a steel where the transformation hardening is increased by an age-hardening process.

This paper will be confined to steels in the third category, but it is worth noting that a considerable amount of research is being carried out on steels in the other two categories, which is also likely to lead to improved properties.



(*a*) Effect of solution treatment on martensite transformation range
(*b*) Effect of primary tempering

Fig. 2—Effect of heat-treatment on martensite transformation range

BACKGROUND TO CONTROLLED-TRANSFORMATION STAINLESS STEELS

Essential features

The main requirements for these types of steel are first, sufficient Cr to make the steels adequately stainless. For oxidation resistance the minimum amount necessary is 12%, and for most of the applications considered 18%Cr should be quite sufficient. It is likely that Cr contents towards the top of this range are preferable. Secondly, a low carbon content is essential to satisfy the requirements for easy fabrication and weldability. The useful range is probably 0·04–0·14%. Thirdly, an essentially austenitic structure is needed. To obtain the maximum properties from the transformation it is desirable that the structure should be almost completely austenitic. Any δ-ferrite which is present will lead to a reduction in the maximum properties and also lead to difficulties in hot working. It will be shown later, however, that a

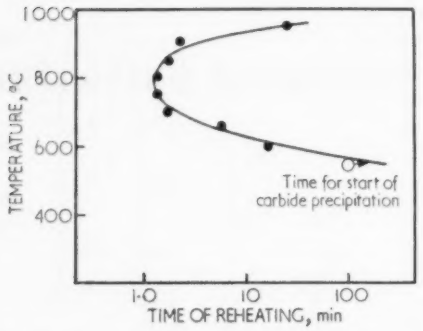


Fig. 3—C-curve for carbide precipitation during reheating of a 17%Cr–4%Ni–3%Mn steel

small amount of δ-ferrite is essential to the production of an all-purpose controlled-transformation steel. Finally, the martensite transformation range should be just below RT. If the M_s is below RT it will mean that an austenitic structure is produced in the solution-treated condition and consequently the steel can be fabricated easily. On the other hand, it means that the martensite transformation range is within reach of a sub-zero treatment or that it can be easily raised by a tempering treatment. Such a steel can also be transformed during cold rolling.

A convenient starting point for the development of these alloys is a 0·1%C–17%Cr composition. Such a steel would be sufficiently stainless, but the ferrite-forming tendency of the Cr requires balancing to produce an essentially austenitic structure. Having obtained the austenitic structure, the martensite transformation range will also need to be controlled. It is well known that the most effective element both as an austenite-former and also in controlling the martensite transformation range is Ni. For instance, it is known that an 18%Cr–8%Ni composition is a very stable austenitic structure which has a martensite transformation range well below RT. Such a structure is far too stable to provide the base for the development of a controlled-transformation stainless steel, and it is necessary to reduce the Ni content. It will be shown later that a 0·1%C–17%Cr–5%Ni composition is still fully austenitic, but the transformation temperature is only just below RT. A 0·1%C–17%Cr–4%Ni composition is just on the borderline of forming δ-ferrite and has its martensite transformation range occurring just above RT. It will be shown later that additional alloying elements are necessary to provide an accurate control of microstructure and transformation temperature, to improve the tempering resistance, and also for age hardening. A 0·1%C–17%Cr–4%Ni composition is very suitable as a base for these further alloy additions.

A detailed description of the effect of each of the different alloying elements is given later, but these can be summarized quite briefly to present the general picture of controlled-transformation steels. The two main effects of these alloying elements are their effect on austenite stability and their effect in depressing the martensite transformation range. These elements fall into two groups which tend either to form ferrite or to form austenite. Typical ferrite-forming elements are Si, Mo, and V, and typical austenite-forming elements are C, Mn, Ni, and Cu. On the other hand, the alloying elements depress the transformation range[3, 4] and therefore it is possible to maintain control of both the structure and the martensite transformation range by using a suitable combination of alloying elements.

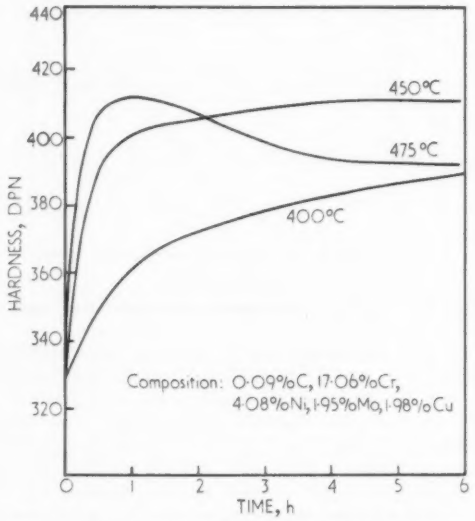


Fig. 4—Precipitation-hardening effect of copper

To illustrate the main features of a controlled-transformation steel two typical compositions have been chosen. These are 0·1%C–17%Cr–4%Ni–2%Mo and 0·1%C–17%Cr–4%Ni–3%Mn. The effects of Mo and Mn on the martensite transformation range are roughly equal but they have opposite effects on ferrite formation. It was mentioned that the 17%Cr–4%Ni base composition was just on the borderline of austenite stability and it was found that 2%Mo introduced 20% of δ-ferrite. In contrast, Mn, which is an austenite former, produces a ferrite-free austenitic structure when added to the 17%Cr–4%Ni base.

Martensite transformation range

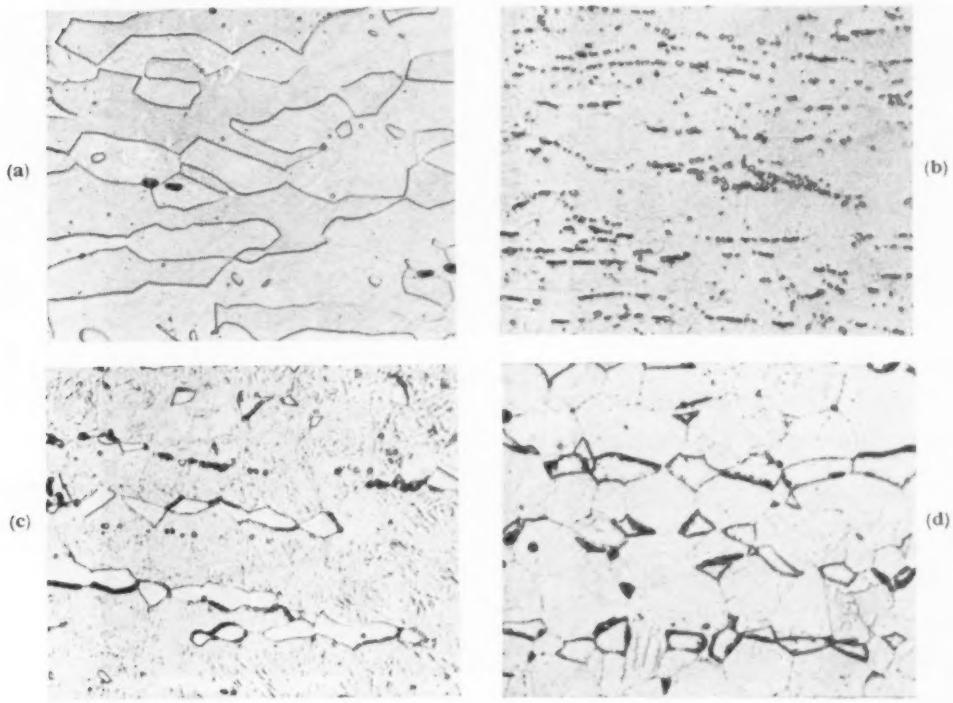
When in the solution-treated condition, the martensite transformation range of both these steels is below RT. Typical figures for the various stages of the martensite transformation are:

	Temperature, °C				
	M_s	M_{10}	M_{50}	M_{90}	M_f
17%Cr–4%Ni–2%Mo	30	−10	−55	−90	−115
17%Cr–4%Ni–3%Mn	−10	−45	−75	−105	−140

These indicate that the martensite transformation range extends over about 140° C. It will also be noticed that the formation of martensite is not uniform over this temperature range, and a typical transformation curve is shown in Fig. 1. The start and especially the finish of the transformation are difficult to determine accurately and these have been indicated by dotted lines.

Effect of solution-treatment temperature

In the description of the martensite transformation range given above, it is assumed that the steel is in



(*a*) 1050° C air cooled
(*b*) 950° C air cooled
(*c*) 1050° C air cooled. Refrigerated — 70° to — 80° C. Tempered 450° C
(*d*) 1050° C air cooled. Primary tempered 700° C

Fig. 5—Typical microstructure of a 17%Cr–4%Ni–2%Mo steel × 750

the completely solution-treated condition. This means that all the carbides have been taken into solution, and therefore both the alloy content and the carbon content of the composition are effective. In simple stainless steels an adequate solution-treatment temperature is 1 050° C. This is the standard temperature used for austenitic stainless steels and for which there is adequate furnace equipment. Higher temperatures create some difficulty and, therefore, compositions which introduce complex carbides needing high solution temperatures must be avoided. On the other hand, there is some benefit in lower solution-treatment temperatures because they enable a control to be exerted on the effective composition.

To produce an effective controlled-transformation steel, the control on the martensite transformation range must be very accurate. On the other hand, to make a steel commercially it is essential to have a steelmaking range for each of the different elements. It will therefore be realized that, allowing for this inevitable variation in alloying elements, there must be some variation in the position of the martensite transformation range. The solution-treatment temperature can be used to control the amount of carbide which is taken into solution, and hence control the effective composition, both with regard to alloying elements and carbon content. The main disadvantage of using a lower solution-treatment temperature is that the out-of-solution carbides reduce the maximum attainable tensile strength and may adversely affect the corrosion resistance. However, provided that the amount of carbide is not too large, it is found that satisfactory properties can be obtained. The effect of solution-treatment temperature on the position of the martensite transformation range is shown in Fig. 2, and it will be seen that a very close control indeed can be obtained. It is perhaps worth stressing here that every cast needs to be treated individually in assessing the correct solution-treatment temperature.

Effect of refrigeration

It is emphasized that, since the martensite transformation is below RT, the steel is austenitic when cooled to RT after solution treatment. In order to transform the steel to a martensitic structure it is necessary to pass through the martensite transformation range. One way in which this can be done is to lower the temperature below the martensite transformation range. The ease with which this can be carried out depends largely on the position and extent of the martensite transformation range. It is relatively simple to obtain commercially a temperature of — 70° to — 80° C using a mixture of solid CO_2 and acetone. Lower temperatures are more difficult to obtain and it is also found that reaction rates become sluggish at very low temperatures, so that transformation rates become extremely slow. Provided that the martensite transformation range is located just below RT, a temperature of — 70° C will bring

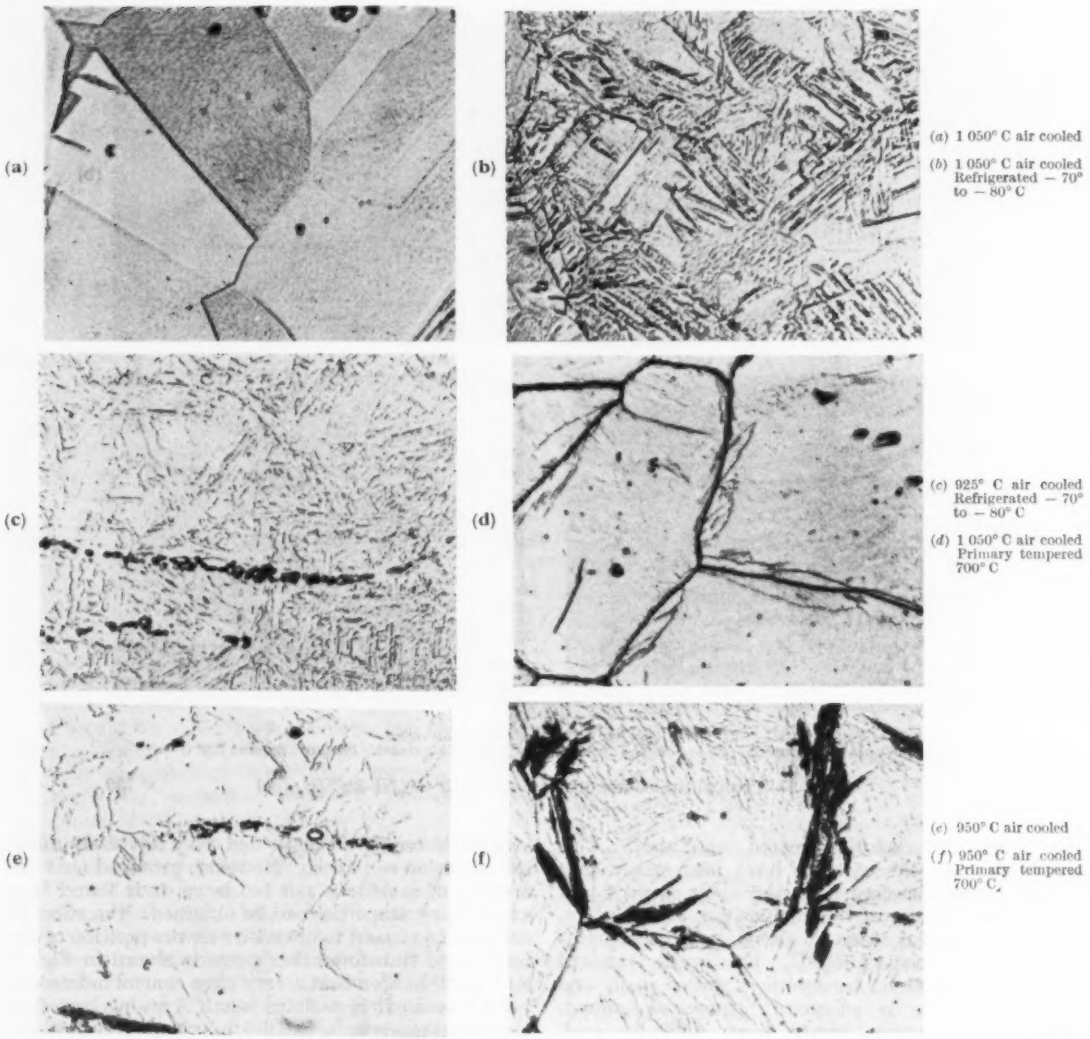

(a)
(b)
(a) 1 050° C air cooled
(b) 1 050° C air cooled Refrigerated − 70° to − 80° C
(c)
(d)
(c) 925° C air cooled Refrigerated − 70° to − 80° C
(d) 1 050° C air cooled Primary tempered 700° C
(e)
(f)
(e) 950° C air cooled
(f) 950° C air cooled Primary tempered 700° C

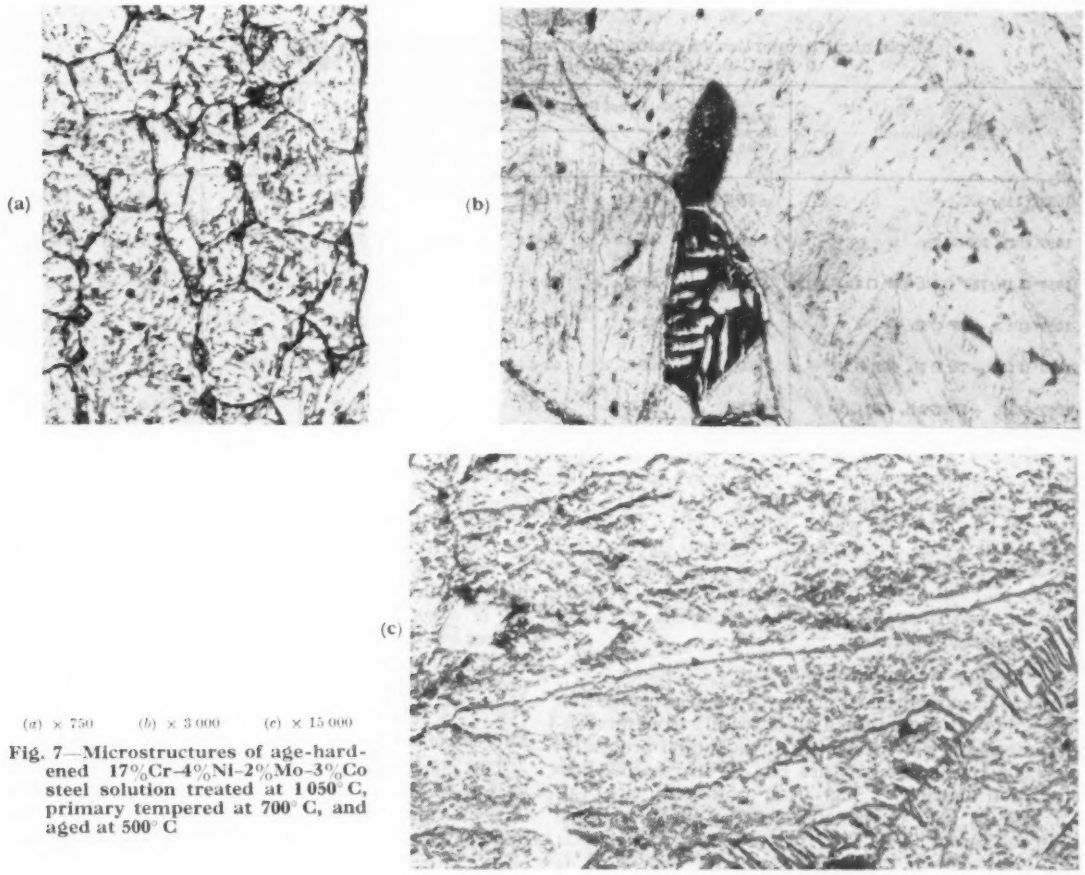
Fig. 6—Typical microstructures of a 17%Cr–4%Ni–3% Mn steel × 750

about practically complete transformation. In the examples quoted in Fig. 2 the amount of transformation will depend on the solution temperature used. If a full solution treatment at 1 050° C is used then − 70° C will produce only 50% transformation. On the other hand, a solution treatment at 950° C raises the martensite transformation range sufficiently for about 95% transformation to occur on cooling to − 70° C. This refrigeration treatment forms the basis of one of the commercial heat-treatments for transforming a controlled-transformation stainless steel.

Effect of primary tempering

In this type of steel, the carbide precipitation reaction is temperature dependent and there is a C-shaped reaction curve. A typical curve is shown in Fig. 3, which indicates that rapid carbide precipitation occurs over a temperature range of 700–800° C. It should be emphasized that this precipitation occurs from austenite and consists of $M_{23}C_6$ (essentially chromium carbide but containing other alloying elements such as Mo or Mn) and Cr_2N. The major precipitate is the carbide, which will remove both carbon and alloying element from the composition. For the heat-treatment to be carried out without scaling or distortion it is preferable to use the lowest temperature which will give rapid precipitation, i.e. 700° C.

The precipitation at 700° C is quite extensive and is sufficient to raise the martensite transformation range by about 160° C. This effect is shown in Fig. 2*b*. It will be seen that the primary tempering treatment has raised the martensite transformation range from being completely below RT to being completely above RT This means that when the steel is cooled to RT from the primary tempering treatment it will now pass through the martensite transformation range and consequently a martensitic structure will be produced. This forms the basis of the second main



(*a*) × 750 (*b*) × 3 000 (*c*) × 15 000

Fig. 7—Microstructures of age-hardened 17%Cr–4%Ni–2%Mo–3%Co steel solution treated at 1 050° C, primary tempered at 700° C, and aged at 500° C

treatment which is applied commercially to controlled-transformation steels. It will be seen that when either a refrigeration treatment or a primary tempering treatment is used the austenitic structure is converted to a martensitic structure. The properties obtained from the refrigeration treatment are always higher than those obtained from the primary tempering treatment, because the carbon content of the primary-tempered structure has been lowered by precipitation. This means that the carbon content of the martensite is lower and consequently its strength is lower.

Effect of secondary tempering

It has been shown above that either of the transformation treatments produces a martensitic structure. It is known that untempered martensitic structures do not possess the best combination of tensile strength and ductility, and it is usually desirable to temper the martensitic structure. When the tempering characteristics of this type of steel are examined it is found that there is relatively little reduction in strength up to a tempering temperature of about 500° C, beyond which there is a rapid softening associated with the decomposition of the martensite and the formation of chromium carbide. Any temperature below this softening point can be used to improve the ductility without reducing the tensile strength. Consequently a secondary tempering treatment at 400–450° C will be used. It is found when different alloys are examined that there are some differences in the tempering characteristics due to the alloying elements used. Generally speaking, this is due to the effect of strong carbide-forming elements, such as Mo, entering into the chromium carbide to retard its formation and to produce a slight secondary hardening process. This effect is much greater with Mo than Mn, and it is found that if a secondary-tempering temperature of 450° C is used there is a hardness increase of roughly 15 DPN in the manganese steel and 40 DPN in the Mo steel. It should be emphasized that the main object of the secondary tempering process in these simple steels is to produce the best combination of strength and ductility rather than to obtain the slightly increased tensile strength. In fact, a slightly higher tempering temperature may be used in order to produce some softening and to obtain still better ductility.

Effect of cold work

It is known that cold work can raise the martensite transformation temperature and, consequently, with controlled-transformation stainless steels it is possible to produce transformation by cold work. Nor-

TABLE I

Mechanical properties of controlled-transformation stainless steel (20% ferrite) 0·10%C–0·21%Si–0·25%Mn–17·32%Cr–4·2%Ni–1·95%Mo

Heat-treatment	Tensile Strength, tons/in²	Proof Stress, tons/in² for extension, %				Limit of proportionality, tons/in²	Elongation on $4\sqrt{A}$, %	Reduction in area, %
		0·20	0·10	0·05	0·02			
1050° C 1h, A.C.	87·0	15·1	14·5	14·1	13·3	9·6	17·8	27·2
1050° C 1h, 700° C 2h	63·5	44·2	36·3	29·5	22·2	11·2	16·8	51·2
1050° C 1h, 700° C 2h, 450° C 1h	69·0	55·3	53·9	52·8	50·0	41·6	18·9	43·2
1050° C 1 h, – 78° C 2h	91·4	52·5	43·8	32·9	22·4	9·6	13·8	27·2
1050° C 1h, – 78° C 2h, 450° C 1h	83·4	64·8	60·3	56·5	51·0	40·0	18·4	47·2
1000° C 1h, – 78° C 2 h, 450° C 1h	79·1	67·0	65·8	64·2	60·3	46·4	20·9	51·2

mally, this effect cannot be relied upon to produce uniform transformation in a fabricated structure, but it could be employed where a flat product or where high-tensile wire is required. Typical mechanical properties of cold-worked steels are given later, but it will be realized that if transformation can be obtained by cold work then the only heat-treatment necessary is a tempering operation.

Precipitation-hardening mechanism

It has been mentioned previously that precipitation hardening can be obtained quite readily from martensitic steels, in which case the already high properties of the martensitic structure can be increased. Precipitation hardening has been obtained with Cu, Co, and Al, and to demonstrate the effect a simple composition of 0·1%C–17%Cr–4%Ni–2%Mo–2%Cu has been chosen. The treatment of this steel is identical to the non-ageing steels mentioned above, and the transformation to a martensitic structure can be obtained either by refrigeration or by primary tempering. In the steels previously referred to, the subsequent secondary tempering operation was mainly to improve the ductility and any hardening was due to the effect of the alloying elements on the decomposition of martensite. In contrast to the small hardness increase obtained during secondary tempering of a non-ageing steel, Fig. 4 shows the hardness curves obtained with the copper-bearing steel. These show that a hardening effect of roughly 80 DPN is now obtained and the curves have the form of a true age-hardening process.

Microstructures

The microstructures of these steels in the various heat-treated conditions are shown in Figs. 5 and 6 and illustrate many of the features already described. The more general case is a steel containing a small amount of δ-ferrite, and Fig. 5 shows structures obtained with a 17%Cr–4%Ni–2%Mo steel. In the 1 050° C solution-treated condition the structure is austenite plus δ-ferrite (Fig. 5*a*). This confirms that the martensite

TABLE II

Mechanical properties of controlled-transformation stainless steel (ferrite free) 0·10%C–17%Cr–4·2%Ni–3%Mn–0·3%Si; ¾-in. dia. rolled bar

Heat-treatment	Tensile Strength, tons/in²	Proof Stress, tons/in² for extension, %				Elongation on $4\sqrt{A}$, %	Reduction in area, %
		0·20	0·10	0·05	0·02		
1050° C	62·5	23·0	21·0	19·2	17·0	60·5	62·4
1050° C, – 78° C 1h, 400° C 2h	85·1	55·4	44·1	34·4	24·8	30·0	46·8
1000° C, – 78° C 1h, 400° C 2 h	85·2	49·3	36·9	27·7	18·8	30·9	49·6
950° C, – 78° C 1h, 400° C 2h	86·8	54·7	46·2	32·6	23·2	30·7	51·4
925° C, – 78° C 1h, 400° C 2h	87·7	73·4	62·5	49·4	33·2	30·5	55·0
1000° C, 700° C 2h, 400° C 2h	83·4	23·3	20·3	18·6	16·5	19·1	45·1
950° C, 700° C 2h, 400° C 2h	78·9	32·8	26·3	21·7	17·4	18·6	46·8
900° C, 700° C 2h, 400° C 2h	74·8	41·3	32·8	24·8	17·6	30·0	52·5
875° C, 700° C 2h, 400° C 2h	73·5	63·5	61·0	58·0	54·8	30·5	54·0

TABLE III

Effect of cold work in producing transformation
0·1%C–17%Cr–4·2%Ni–3%Mn–0·3%Si; 0·2-in. thick sheet; 925° C, cold work, 450° C 3 h

Reduction by Rolling, %	Tensile Strength, tons/in²	Proof Stress, tons/in² for extension, %						Limit of Proportionality, tons/in²	Elongation on 2 in., %
		0·50	0·30	0·20	0·10	0·05	0·02		
0	62·0	50·8	44·5	39·5	31·6	25·9	20·4	10·0	10·0
5	60·8	52·4	48·4	42·8	32·8	25·6	17·4	7·0	6·0
10	79·3	...	65·0	61·1	50·9	40·4	31·0	19·5	8·5
15	86·6	...	...	83·7	78·7	72·8	61·4	45·5	8·5
20	87·2	...	...	84·5	81·1	77·0	69·9	53·0	4·8
25	97·7	...	...	96·6	92·7	89·0	81·5	65·5	3·2
30	101·8	...	...	99·2	97·8	92·7	74·0	72·8	4·2
40	110·0	...	...	109·8	106·2	101·0	91·2	63·7	3·5

transformation range is completely below RT. A lower solution temperature of 950° C leaves an appreciable amount of carbide out of solution, and, because the martensite transformation range has been raised, some of the austenite has transformed to martensite (Fig. 5*b*). When refrigeration is applied, the austenite transforms to martensite, but the structure is very light etching until it is tempered (Fig. 5*c*). A primary tempering treatment at 700° C produces appreciable carbide precipitation which occurs mainly at the δ-ferrite boundaries, and since this raises the martensite transformation range the matrix has transformed to martensite on cooling to RT (Fig. 5*d*). Once again, this martensite does not show clearly until secondary tempering is carried out.

The other type of structure encountered is one which is ferrite free, and a typical steel, 17%Cr–4% Ni–3%Mn, is illustrated in Fig. 6. The completely austenitic structure after solution treatment (Fig. 6*a*) is transformed to martensite by refrigeration (Fig. 6*b*). This is not complete, however, because the martensite transformation range is rather low. A solution treatment at 925° C corrects this and complete transformation can then be produced by refrigeration (Fig. 6*c*). It has been found that this steel is resistant to transformation by primary tempering and the microstructures show that this is due to the limited extent of the carbide precipitation. After primary tempering there is a small amount of precipitation around austenite grain boundaries and adjacent to this some martensite has formed (Fig. 6*d*), but the amount of carbide precipitation is not enough to raise the transformation range sufficiently for full transformation to occur. It is found that lower solution temperatures can make the steel respond to primary tempering and this feature can also be explained by the microstructure. The lower solution temperature produces some martensite in the microstructure (Fig. 6*e*). (Fig. 2*a* shows how the martensite transformation range is raised above RT) and this tempers rapidly during the 700° C treatment (Fig. 6*f*). Heavy carbide precipitation accompanies this tempering process, and the transformation range is raised and martensite formation produced. The dark etched regions in the microstructure are the martensite areas which existed before the primary tempering.

With ageing compositions the only effect which can be seen in the optical microscope is a darkening of the microstructure during primary tempering (Fig. 7*a*). Examination at higher magnification using the electron microscope confirms the presence of a precipitate (Figs. 7*b* and *c*).

Typical mechanical properties

The foregoing discussion has shown how different heat-treatments affect the transformation behaviour of these steels. Typical mechanical properties of the two steels mentioned were determined and these are given in Tables I and II. The results obtained illustrate the essential feature of these steels, namely, that heat-treatment produces transformation to martensite.

It may seem surprising that tensile strengths as high as 80 tons/in² are obtained from these steels in the austenitic condition. This is not a true reflection of the strength however, since it results from the extensive transformation which occurs during necking before fracture of the tensile specimen. A better indication is obtained from the hardness, which is about 200 DPN. A more important property is the 0·2% proof stress and it will be seen that this is low before transformation and is typical of an austenitic steel.

The 17%Cr–4%Ni–2%Mo steel (which contains 20% of ferrite), responds readily to both refrigeration and primary tempering. The high tensile strength and high PS/TS ratio obtained is an indication that complete transformation to martensite has taken place. These mechanical properties show the difference between refrigerated and primary-tempered conditions and it will be seen that there is a difference of some 10 tons/in² in the tensile properties. This is a reflection of the lower carbon content of the martensite in the primary tempered condition.

TABLE IV

Effect of tempering cold-rolled sheet
0·1%C–17%Cr–4·2%Ni–3%Mn–0·3%Si; 0·045-in. thick

Condition	Tensile Strength, tons/in²	Proof Stress, tons/in² for extension, %						Limit of Proportionality, tons/in²	Elongation on 2 in., %
		0·50	0·30	0·20	0·10	0·05	0·02		
As-rolled	98·6	67·6	60·0	53·4	43·7	35·2	27·4	14·5	5·8
400° C ½ h	94·5	...	75·2	68·5	56·0	47·4	38·8	21·5	10·8
400° C 1 h	94·4	...	79·1	72·3	59·1	47·1	35·5	17·0	7·0
400° C 2 h	94·2	...	78·4	70·9	58·0	46·5	32·1	16·5	14·3
450° C ½ h	89·5	83·8	76·8	69·5	54·5	44·0	30·6	17·0	5·0
450° C 1 h	97·4	...	78·5	73·0	61·5	49·8	39·0	22·0	6·1
450° C 2 h	91·4	78·9	75·0	69·3	58·7	48·0	38·3	23·0	14·6
500° C ½ h	85·6	...	...	69·0	58·8	47·6	37·6	26·6	15·5
− 78° C 1 h, 400° C 2 h	93·0	...	...	77·0	66·9	57·6	47·6	32·4	5·6

The 17%Cr–4%Ni–3%Mn steel responds readily to refrigeration to give similar properties to the 2%Mo steel. It does not respond so easily to primary tempering because of the absence of δ-ferrite. To produce transformation a low solution temperature must be used, which forms a small amount of martensite in the solution-treated condition. The microstructure in Fig. 6*f* shows how this subsequently tempers during the primary tempering treatment.

One of the features in early trials of controlled-transformation steels was the variable properties obtained. Whilst the maximum properties were satisfactory, considerably lower properties were frequently obtained, indicating that transformation had not been complete. This effect of the amount of transformation can be seen quite easily using a controlled-transformation steel which has a martensite transformation range below RT and which is then refrigerated at various temperatures between 0° and − 80° C. This has produced a series of steels containing various amounts of martensite and the properties obtained are instructive. Typical properties of bar material refrigerated and tempered for 2 h at 400° C are given below:

		Martensite, %					
		0	20	47	65	82	100
Tensile strength, tons/in²		78·4	82·3	80·0	78·3	84·6	88·2
Proof stress, tons/in² for extension, %	0·2	14·2	14·2	18·4	36·2	47·6	63·8
	0·1	12·4	13·5	15·8	28·0	37·1	51·8
	0·05	11·2	12·6	13·8	21·4	28·9	38·4
	0·02	9·8	11·2	12·1	15·2	21·5	23·6
Elongation on $4\sqrt{A}$, %		42·4	37·7	44·5	30·0	33·9	27·9
Reduction in area, %		53·8	40·0	51·2	54·4	48·8	55·0

It will be seen that there is little change in the most important property, the 0·2% PS value, until there is more than 50% martensite present. This proof stress value then gradually increases with increasing amounts of martensite, and satisfactory properties

TABLE V

Mechanical properties of a controlled-transformation stainless steel
0·09%C–0·69%Mn–0·33%Si–17·06%Cr–4·08%Ni–1·95%Mo–1·98%Cu; ¾-in. dia. rolled bar

Heat-treatment	Tensile Strength, tons/in²	Proof Stress, tons/in² for extension, %				Limit of Proportionality, tons/in²	Elongation on $4\sqrt{A}$, %	Reduction in Area, %
		0·20	0·10	0·05	0·02			
1000° C 1h A.C.	71·4	16·3	15·4	14·4	12·0	8·0	22·6	43·2
1000° C A.C., − 78° C 1h, 400° C 6h	88·6	70·6	67·2	57·6	57·2	46·8	17·1	43·2
100° C A.C., − 78° C 1h, 450° C 6h	87·0	76·2	72·1	66·6	59·2	44·8	17·1	40·0
1050° C 1h A.C.	84·1	14·2	13·0	11·0	8·4	3·2	31·1	47·2
1050° C A.C., 700° C 2 h, A.C.	82·4	43·2	34·4	25·4	17·6	10·4	13·0	40·0
1050° C A.C., 700° C 2 h, 450° C 1h	80·2	65·7	63·4	61·4	56·6	34·4	14·7	40·0
1050° C A.C., 700° C 2h, 450° C 6 h	78·4	67·7	66·1	63·4	62·4	32·0	12·5	36·0
1050° C A.C., 700° C 2 h, 450° C 24h	81·8	74·4	71·9	70·7	62·4	37·6	9·4	16·4

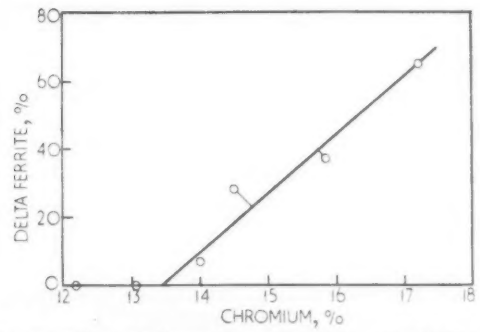


Fig. 8—Effect of chromium content on constitution of 0·1% carbon steels solution treated at 1 050° C

are not obtained until practically complete transforma- to martensite has occurred. It is quite clear, therefore, that the reason for the variability in proof-stress values obtained is due to incomplete transformation to martensite over the range 80–100% transformation. Further tests were carried out to show this effect more clearly, and in these steels it was arranged that various amounts of untransformed austenite should be present, varying from 18% to 0%. The tensile properties obtained on bar material refrigerated and tempered for 2 h at 400° C are given below:

Property		Estimated Residual Austenite, % 18	13	8	5	0
Tensile strength, tons/in²		84·6	85·1	84·7	85·4	88·2
Proof stress, tons/in² for extension, %	0·2	47·6	55·4	62·4	63·9	63·8
	0·1	37·1	44·1	50·7	54·7	51·8
	0·05	28·9	34·4	41·2	44·0	38·4
	0·02	21·5	24·8	31·0	30·6	23·6
Elongation on $4\sqrt{A}$, %		33·9	30·0	29·4	25·3	27·9
Reduction in area, %		48·8	46·8	52·9	43·2	55·0

It will be seen that to ensure a satisfactory PS/TS ratio, it is necessary to achieve more than 90% transformation to martensite.

This raises the interesting point with regard to the correct control of the martensite transformation range, which applies when refrigeration is to be used for transformation. It has been shown that the martensite transformation range is some 140° C and it has also been stated that the transformation range should occur between RT and — 70° to — 80° C. It will be realized that the martensite transformation range must overlap this range either above or below. It is necessary therefore to decide which of these two alternatives is preferable. When the properties given above are considered it will be realized that it is better to have the martensite range extending slightly above RT rather than to have it extending below — 80° C. This is because small amounts of martensite present in the solution-treated condition will still permit fabrication of the steel, although, since transformation is occurring during deformation, the fabrication properties cannot equal those of a completely stable austenite. On the other hand, small amounts of residual austenite following refrigeration will lower the properties obtainable.

There is one other interesting feature of incomplete transformation which may well affect the steel after primary tempering. Once again, if the treatment has not been completely effective, the final part of the martensite transformation range may extend below RT. Consequently, when air cooled, lower mechanical properties are obtained. It is possible to rectify this by applying a partial refrigeration treatment, i.e. at temperatures of the order of — 10° or — 25° C. This will enable the final transformation to be carried out; such treatment is possible with much existing refrigeration equipment.

It is known that cold work will raise the martensite transformation range, and consequently this provides one means of transforming these steels. Mechanical properties are given in Table III which show how increasing amounts of cold work produce a breakdown of the austenitic structure and then subsequently, cold work this structure. Roughly 15%–20% reduction is needed to produce transformation. After tempering quite satisfactory mechanical properties can be obtained. If this method of transformation is to be

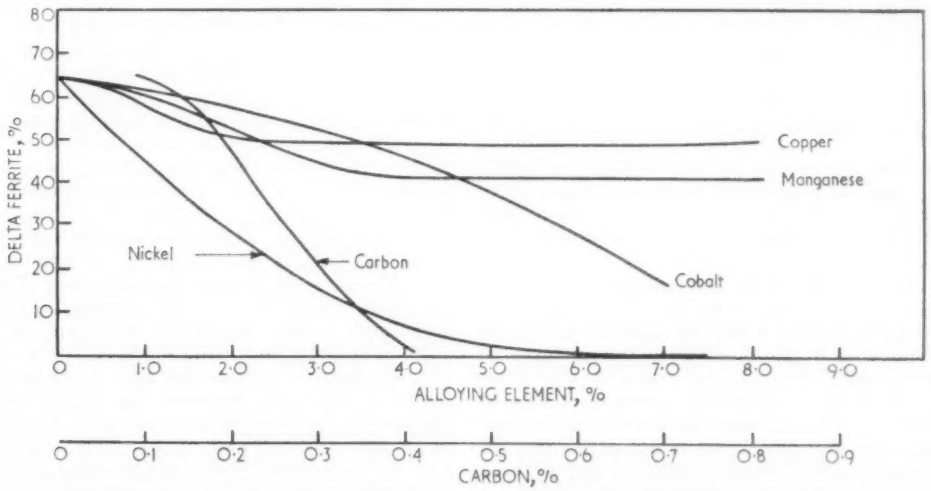


Fig. 9—Effect of various alloying elements on constitution of 0·1%C–17%Cr alloys

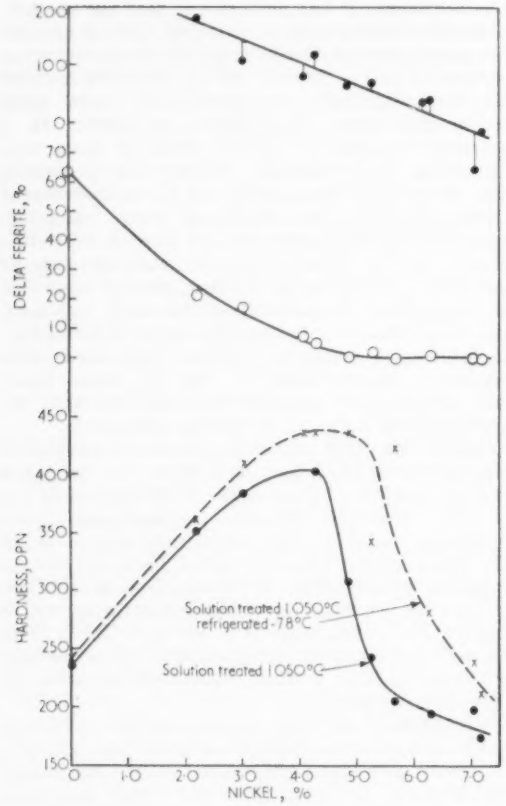


Fig. 10—Effect of nickel content on constitution, hardness, and M_s temperature of 0·1%C–17%Cr steels

used, then the composition can be specially adjusted to give transformation after the appropriate amount of cold work.

In Table IV properties after tempering are given for commercially produced cold-rolled sheet with a final reduction of about 20%. Whilst the mechanical properties are quite satisfactory, a duplicate test using refrigeration showed that transformation had not been complete during rolling. To accomplish this a slightly increased amount of cold work should be used.

It has been shown that Cu will produce an ageing reaction and consequently increased mechanical properties would be expected. The 17%Cr–4%Ni–2%Cu–2%Mo was tested and the mechanical properties are given in Table V.

It will be seen that these mechanical properties are higher than those obtained in the copper-free steels and the difference is due to the precipitation hardening of the copper-rich phase. Whilst the increase in properties is useful, the most valuable feature of these results is that very satisfactory properties can be obtained even when the transformation to martensite during the previous treatment is not complete. An example is given in the mechanical properties which shows that after primary tempering at 700° C the 0·2% PS value was only 43 tons/in², indicating

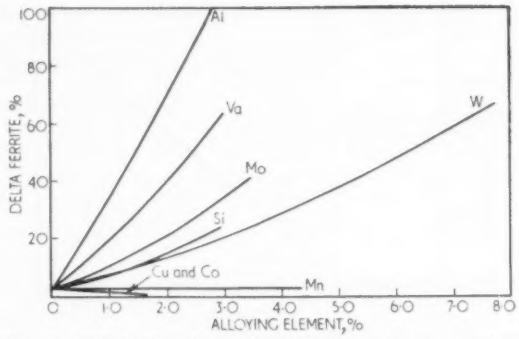


Fig. 11—Effect of various alloying elements on constitution of 17%Cr–4%Ni alloys

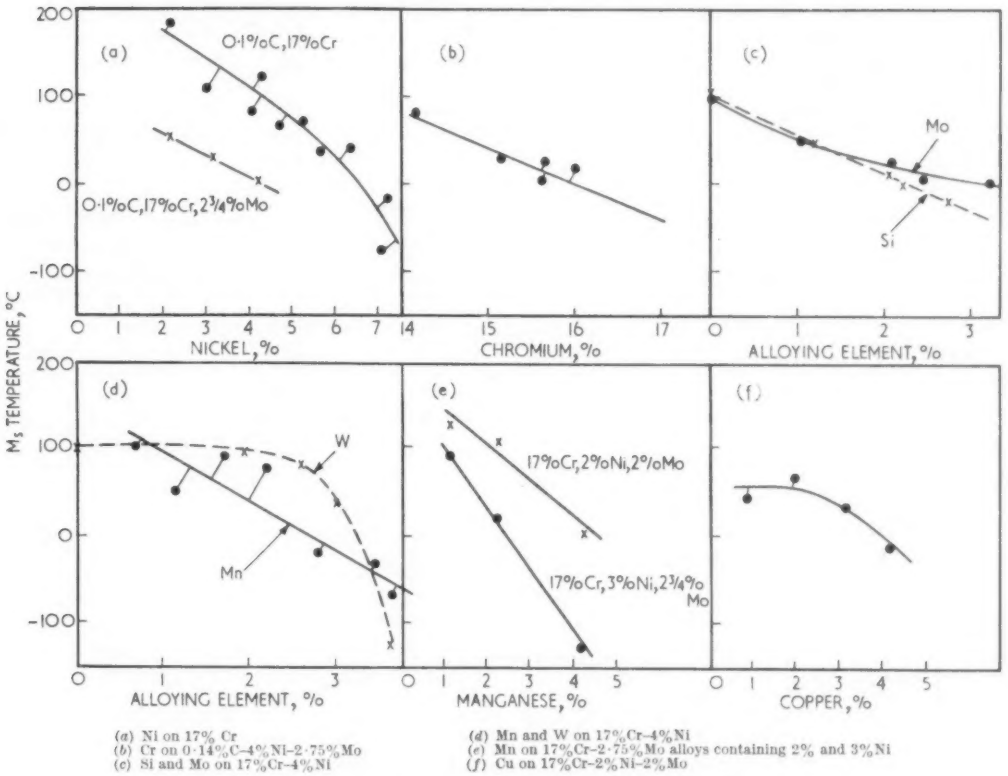
incomplete transformation to martensite. The secondary hardening treatment has raised this to 70 tons/in², which is very satisfactory. It has been concluded from these results that, whilst quite satisfactory properties can be obtained from simple controlled-transformation steels, they are extremely difficult to control and any departure from complete control results in considerably poorer mechanical properties. The secondary-hardening steels not only offer a possibility of slightly higher maximum properties, but they are much more flexible and can produce high properties, even when the control of the martensite transformation is not completely accurate. They are more likely therefore, to be acceptable as commercial steels.

EXPERIMENTAL WORK ON CONTROLLED-TRANSFORMATION STEELS

Choice of base composition

The starting point for a controlled-transformation steel is a low carbon content (0·1%) and sufficient Cr for adequate corrosion resistance. The two features which must continually be kept in mind are the structure and the martensite transformation range. The effect of increasing Cr (above the absolute minimum for oxidation resistance) on the structure formed at 1 050° C is shown in Fig. 8. Ferrite is introduced into a 0·1%C steel when the Cr content is raised above 13½%, and with 17%Cr there is 64% of δ-ferrite. A 0·1%C–17%Cr steel also has an M_s point at 250° C; consequently, to produce a controlled transformation steel from this base composition, it is necessary to introduce alloying elements which will reduce the ferrite content and lower the martensite transformation range. The elements which reduce ferrites are Ni, Cu, Mn, Co, C, and N and each of these has been studied. A comparison of the effects of these elements is shown in Fig. 9, and it will be seen that Ni is by far the most effective. This is the only element which is capable of reducing the ferrite sufficiently without depressing the M_s too far.

The two most important effects of Ni are shown in Fig. 10, where it will be seen that 4%Ni reduces the ferrite to 5% and depresses the M_s to 100° C. A further 1%Ni eliminates ferrite and depresses the M_s almost below RT so that a 17%Cr–5%Ni is in fact a controlled-transformation steel. There are some disadvantages in this composition however, because the analysis range is quite critical. It will be



(*a*) Ni on 17% Cr
(*b*) Cr on 0·14%C–4%Ni–2·75%Mo
(*c*) Si and Mo on 17%Cr–4%Ni
(*d*) Mn and W on 17%Cr–4%Ni
(*e*) Mn on 17%Cr–2·75%Mo alloys containing 2% and 3%Ni
(*f*) Cu on 17%Cr–2%Ni–2%Mo

Fig. 12—Effects of alloying elements on M_s temperature

seen in Fig. 10 that low Ni contents lead to low maximum hardness values in the transformed condition because of the large amount of δ-ferrite present. High Ni contents produce a steel which is difficult to transform by primary tempering because of the absence of ferrite, and which is also difficult to transform by refrigeration because of the marked depression of the martensite transformation range. It is for this reason that a 17%Cr–4%Ni composition was chosen as the base, and the final control of ferrite content and the position of the martensite transformation range is accomplished by small additions of other alloying elements.

Effect of alloying elements on a 17%Cr–4%Ni base composition

With the base composition established, the next step was to examine the effect of other alloying elements added to this base. The effects of various ferrite-forming elements are shown in Fig. 11. It will be seen that the ascending order of effectiveness in forming δ-ferrite is W, Si, Mo, V, and Al. This diagram shows also the effect of austenite-forming elements (Mn, Cu, and Co), but since the base composition contains only 5% of ferrite, this does not show the true effect of these elements in steels containing larger amounts of ferrite. This effect can be seen more clearly in Fig. 9.

It would be very convenient if, as a result of work of this type, the ferrite-forming tendencies of various elements could be established and applied to further alloy development. It is not possible to do this accurately, because in many cases the effect is not linear; nor is the mean effect the same with different base compositions. However, by examining a large number of different steels it is possible to obtain a general estimate which is extremely useful in planning experimental work. Tests carried out on a relatively small number of steels will then make it possible to select the optimum composition.

The average values for the ferrite-forming effects of the various alloying elements which have been derived from a study of many different base compositions are as follows:

Element	Change in δ-ferrite for 0·1 wt-%, %	for 1 wt-%, %
N	− 20	...
C	− 18	...
Ni	...	− 10
Co	...	− 6
Cu	...	− 3
Mn	...	− 1
W	...	+ 8
Si	...	+ 8
Mo	...	+ 11
Cr	...	+ 15
V	...	+ 19
Al	...	+ 38

Elements such as Ti and Nb have not been assessed in this manner because, in addition to affecting ferrite, they also remove carbon from solution and thereby promote further ferrite formation. This effect may

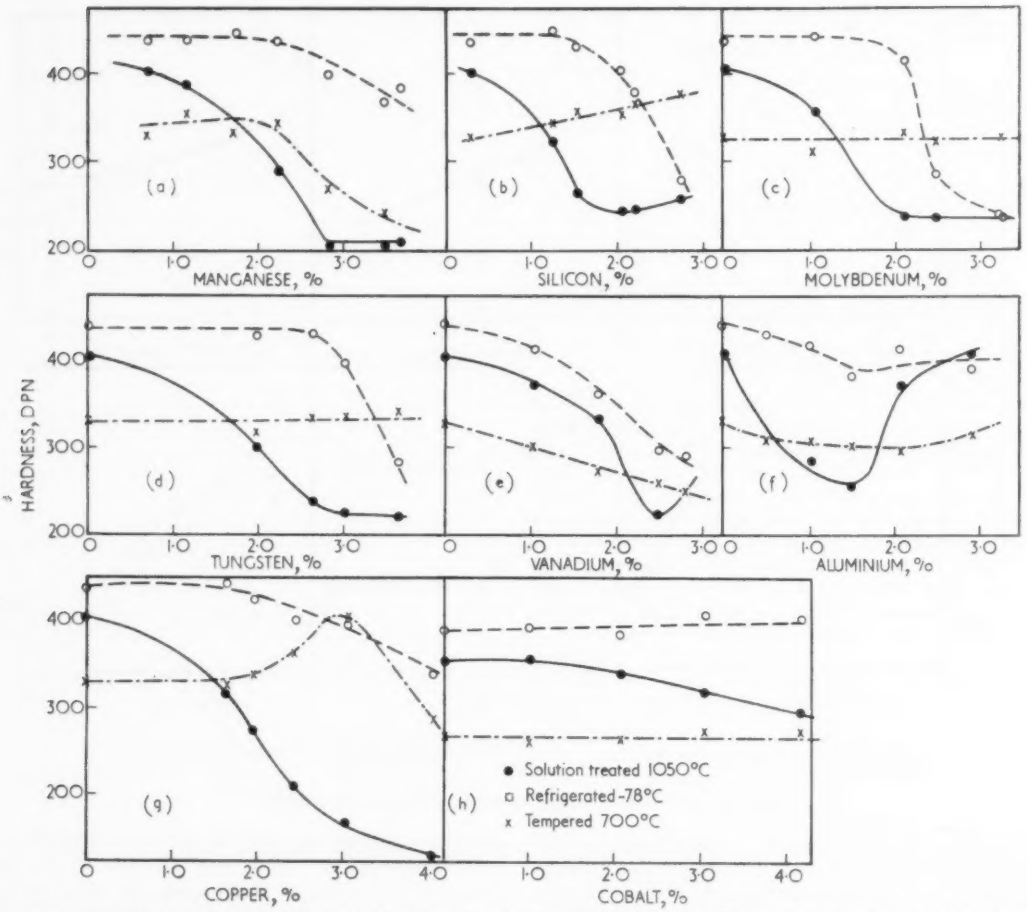


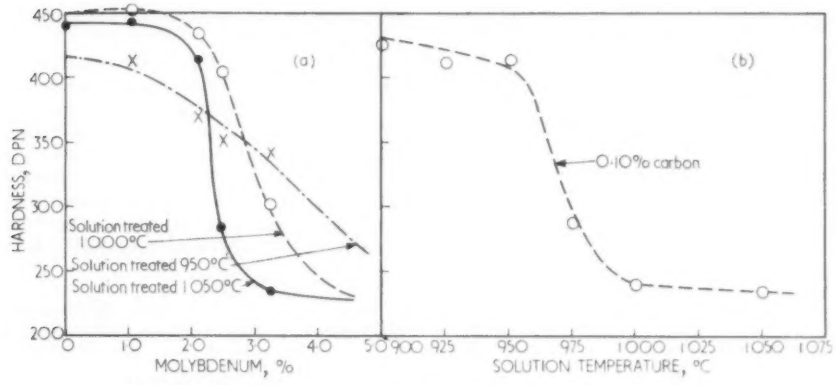
Fig. 13—Effect of various alloying elements on the reaction to heat-treatment of a 17%Cr–4%Ni alloy

occur to a lesser extent with V and also with Al (owing to combination with nitrogen). Since nitrogen has such a pronounced effect it is essential that it is closely controlled in steelmaking. In this investigation nitrogen has been controlled at 0·02%, unless intentionally high nitrogen contents were required.

The other feature studied of these relatively simple compositions was the depression of the martensite transformation range. In the early stages of this work, the effect of each alloying element was studied dilatometrically. Typical results from this work are shown in Fig. 12. The effect of Ni and Cr is shown in Figs. 12*a* and *b*, and it will be seen that Ni has a slightly different effect according to the base used. The effect of several alloying elements (Mo, Si, W, and Mn) on a 17%Cr–4%Ni base is shown in Figs. 12*c* and *d*. In the case of tungsten the effect is definitely non-linear. The graphs show that each of these elements is capable of producing a controlled-transformation steel on this base composition. Figures 12*e* and *f* show how other austenite-forming elements can be used to produce controlled transformations with lower Ni contents in the base composition.

During this work it was realized that the real test of a controlled-transformation steel was the amount of martensite present after the various heat-treatments. This could easily be checked by hardness measurements and, in addition to being much simpler, was also more representative of actual heat-treatment conditions than a small sample cooled at a different cooling rate in a dilatometer. In studying more complex steels, therefore, hardness testing has generally been used, although these results have been checked dilatometrically, particularly when information was required about the extent of the transformation range.

The complex picture is obtained by comparing the effect of alloying elements on δ-ferrite formation and in depressing M_s. The combination required is a martensite transformation range just below RT with a ferrite content of 8–20% (to enable maximum properties to be obtained with ready response to primary tempering). The base composition of 17%Cr–4%Ni is practically austenitic with an M_s of 100° C. The figures given below show the amount of alloy which is required to depress the M_s below RT and the amount of ferrite which would be produced.



(*a*) Mo on 17%Cr–4%Ni (*b*) C on 17%Cr–4%Ni–2·75%Mo–0·1%N_2 alloys

Fig. 14—Effect of solution treatment on response of alloys to refrigeration hardening

Element	Amount necessary to depress M_s, %	Ferrite produced, %
Mo	2	22
Si	2	16
W	3	24
Mn	3	0

If in developing these steels there is any tendency for the amount of ferrite to be too high, it is possible to modify this by substituting austenite-forming elements. Since, however, apart from Ni these have a weak effect, the modification is best made to the base composition. If the combined effects of Cr and Ni are considered it will be seen that 16%Cr–5%Ni and 15%Cr–7%Ni compositions have similar M_s points, but have increasing ability to counteract additions of strong ferrite-forming elements.

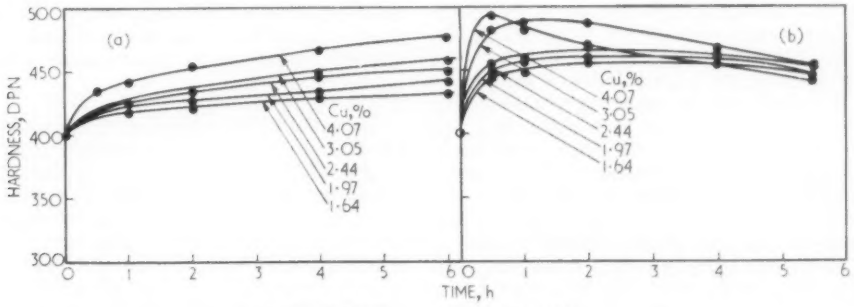
Effect of alloying elements on controlled-transformation characteristics

It has been shown in the introduction that the main features of a controlled-transformation steel are that it has an austenitic structure after solution treatment, but can be hardened either by refrigeration or primary tempering. These transformation characteristics can be studied quite easily by means of hardness measurements in the solution-treated, refrigerated, and primary-tempered conditions. Much information can be obtained from these simple tests and this can then be supplemented by tensile properties on selected steels to give further information on PS/TS ratio and ductility values. The simple hardness tests, however, indicate whether the main features of a controlled-transformation steel have been obtained. This can be demonstrated by considering the effects of the important alloying elements, Mn, Si, Mo, W, V, Al, Cu, and Co, on a 17%Cr–4%Ni base, and these are shown in Fig. 13. The important features of these results are summarized in the succeeding paragraphs.

(1) Since the base composition has a martensite transformation range completely above RT, it is completely martensitic in the solution-treated condition. The alloying elements all depress M_s and the consequent decrease in the amount of martensite formed is shown by the decreasing hardness of the solution-treated steel. When the hardness value drops to about 200 DPN this is an indication that the M_s is below RT and the structure contains no martensite. It will be seen that this effect is shown by all the alloying elements except Co, which means that Co has a very small effect in depressing M_s.

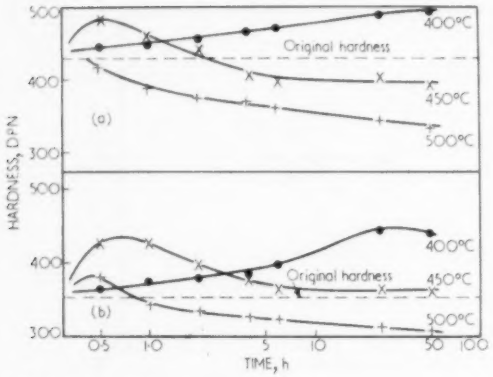
(2) When the steels contain a large amount of ferrite in the solution-treated condition, it will be seen that there is an increase in hardness with increasing alloy content. This applies to Si, V and Al (Figs. 13*b*, *e*, *f*) and is presumably due to a greater concentration of alloying element in the ferrite.

(3) It will be seen that the hardness curve in the refrigerated condition follows the same trend as the solution-treated curve; this is to be expected since it is simply a measure of the effect of the alloying element



(*a*) Aged at 400° C (*b*) Aged at 450° C

Fig. 15—Effect of Cu content on secondary hardening of a 17%Cr–4%Ni alloy. Initial condition: solution treated at 950° C, refrigerated at − 78° C



(*a*) Initial condition: refrigerated — 78° C
(*b*) Initial condition: tempered 700° C

Fig. 16—Secondary hardening of a 17%Cr–3%Ni–3%Mn alloy containing 2%Cu

in depressing the M_s below — 78° C. From these results it is possible to get a rough assessment of the effect of the various alloying elements in depressing M_s. In the table below, the amount of alloy necessary to depress the M_s from 100° to 20° C and from 100° to — 80° C is given. It will be seen that the effect is not linear, but it increases markedly as the alloy content increases. As with the effect of alloys on ferrite formation, a rough estimate of the effect in depressing M_s is useful in planning experimental steels:

Element	Alloy to depress M_s from 100° to 20° C, %	Average depression of M_s, ° C/1%	Alloy to depress M_s from 100° C to — 80° C, %	Average depression of M_s, ° C/1%
Mn	3	27	5	35
Cu	2½	32	4½	40
W	3	27	4	45
Mo	2	40	3½	51
Al	1½	54	3½	51
Si	2	40	3	60
V	2½	32	3	60

The average value obtained over the range 100–20° C is probably the more useful in developing controlled-transformation steels.

(4) The mechanism of the 700° C primary tempering treatment has been discussed in the introduction, and Fig. 13 shows that all steels respond in some degree to this heat-treatment. Generally a fairly uniform hardness is obtained which is lower than the maximum obtained by refrigeration. This is because the 700° C treatment involves the precipitation of carbides, which lowers the carbon content of the martensite finally produced. It will be seen that the Mn steels are resistant to this primary tempering when there is no ferrite in the microstructure, and this has been explained as being due to a lack of suitable precipitation sites. Where the solution-treated hardness is high owing to the formation of martensite, primary tempering merely results in softening.

The carbide precipitation which occurs during the 700° C treatment can have such a marked effect in raising the martensite transformation range that it is possible to harden a steel which will not respond to a — 78° C refrigeration. Consequently, if primary tempering is to be adopted as the commercial heat-treatment, it is possible to allow slightly greater variation in steel analysis and still obtain satisfactory properties.

(5) The effect of various solution treatments in varying the effective composition has been mentioned in the introduction. Figure 14 shows this effect for 17%Cr–4%Ni–Mo steels on the response to a refrigeration treatment. At the lower temperatures it is possible to transform higher alloy compositions, although the maximum hardness obtainable drops because of the amount of carbon which has been removed as a result of carbides being out of solution.

Secondary hardening

In all controlled transformation steels the martensitic structure must be tempered to give good ductility values. If a temperature of 400–450° C is used, this can be carried out without loss of strength and in fact, a slight increase can be obtained. Molybdenum is the most effective element in giving enhanced tempering resistance and in producing a slight age-hardening reaction (it is suggested this is because it enters into the chromium carbide produced on tempering and causes extra coherency straining of the matrix).

In addition to this effect, it is possible to obtain marked secondary hardening using certain alloying elements (Cu, Co, Al, Ti, Nb). These elements form definite age-hardening systems and it has been found that the precipitation reactions are more effective from a martensitic structure in which they occur over the temperature range 400–500° C. The precipitation treatment can therefore be carried out as a secondary tempering operation following either refrigeration or primary tempering. At the present time, the basic mechanisms involved have not been studied, nor have the precipitating phases been identified, although precipitates have been clearly seen (Fig. 7*c*). Work is now being carried out on this aspect and will be reported in a later paper. It seems likely that some of these precipitates are intermetallic compounds.

Copper

Secondary hardening reactions due to Cu have been studied in several different steels. Figure 4 has shown the effect in a 17%Cr–4%Ni–2%Mo–2%Cu steel and similar effects are obtained in other steels. Figure 15 shows that higher Cu contents produce an increased hardening effect. Figure 16 shows the effect of secondary hardening after primary tempering, indicating that adequate response to age hardening is obtained. The main feature of these Cu steels is that they produce quite significant secondary hardening, but over-ageing occurs rapidly above 450° C.

Cobalt

There is fairly strong evidence that Co is not effective in promoting secondary hardening in a 17%Cr–4%Ni steel although it does increase the tempering resistance. Figure 17 and 18, however, show that cobalt has a marked secondary hardening effect in a 17%Cr–4%Ni–2%Mo steel. The hardening effect increases with increasing cobalt, and there is no indication of over-ageing after heating for 50 h at 500° C. The dependence of the reaction on the presence of Mo is perhaps evidence of a particular intermetallic compound.

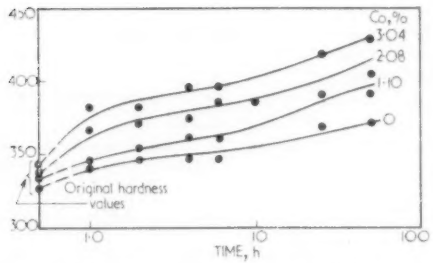


Fig. 17—Effect of varying Co content on secondary hardening of 17%Cr–4%Ni–2%Mo alloys aged at 450° C. Initial condition: solution treated at 1 050° C, tempered at 700° C

Aluminium

Aluminium has perhaps the strongest secondary hardening effect of any element studied. Figure 19 shows that hardness increases of 150–200 DPN can be obtained in a 17%Cr–4%Ni–3%Mn steel after ageing at 450° C. At this temperature the hardness is still increasing after 100 h, but over-ageing and softening occurs quite rapidly at 500° C. Secondary hardening occurs in both the refrigerated and primary-tempered conditions and in fact, an unusually higher hardness was obtained after the primary tempering treatment. Similar hardening effects with Al have been obtained with other base compositions.

Titanium and niobium

A limited amount of work has shown that secondary hardening of 100–150 DPN can be obtained with tempering at 450° C but over-ageing occurs at 500° C.

Summary of the effects of different alloying elements

Based on the results which have been given in previous sections, the main effects of alloying elements in controlled transformation steels are summarized in the following paragraphs.

Chromium—This is one of the essential elements for providing a base composition with adequate corrosion resistance. Whilst 17%Cr is preferred, lower Cr contents give more freedom in developing alloy compositions, since larger amounts of ferrite-forming elements can be used.

Nickel—The other essential element in the base composition, since high δ-ferrite contents can be eliminated without depressing the M_s too far below RT, 17%Cr–4%Ni is a suitable composition.

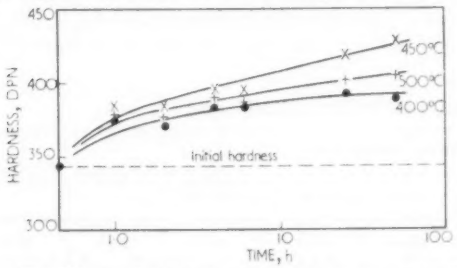
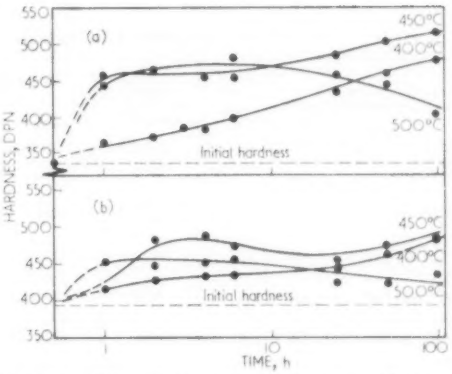


Fig. 18—Effect of ageing temperature on secondary hardening of a 17%Cr–4%Ni–2%Mo–3%Co alloy. Initial condition: solution treated at 1 050° C, tempered at 700° C



(*a*) Initial condition: solution treated 1 050° C, tempered 700° C
(*b*) Initial condition: solution treated 1 050° C, refrigerated — 78° C

Fig. 19—Effect of 1%Al on secondary-hardening characteristics of a 17%Cr–4%Ni–3%Mn alloy

Carbon—Carbon is very effective in depressing M_s and is also an austenite-forming element. If higher carbon contents are used, the total alloy content must be reduced to retain control of the transformation. Since Ni is required for austenite stability, this reduction must be made in the Cr content. With higher carbon contents, higher tensile strengths are possible, but this must be weighed against lower corrosion resistance and possible welding difficulties. Because of this it seems preferable to keep carbon low (0·1% maximum) and to obtain high strengths by an ageing reaction.

Nitrogen—This element has a very similar effect to carbon, but will probably give less corrosion trouble. Whilst it may, therefore, be used with advantage to produce high strengths, it may be difficult to control in steelmaking.

Manganese—A cheap alloying element which will provide a controlled transformation steel at 17%Cr–4%Ni–3%Mn. Since Mn is an austenite-forming element, this composition is ferrite-free and has excellent hot- and cold-working characteristics. On the other hand, the absence of ferrite makes the steel difficult to transform by primary tempering.

Molybdenum—Molybdenum is a ferrite-forming element which will give a controlled-transformation steel with 20% δ-ferrite at 17%Cr–4%Ni–2%Mo. This composition responds readily to both primary tempering and refrigeration, and the Mo has the effect of increasing the tempering resistance and improving the high-temperature properties.

Silicon—Silicon is a cheap ferrite-forming element which will give a controlled-transformation steel with 16% δ-ferrite at 17%Cr–4%Ni–2%Si. This steel will respond readily to both primary tempering and refrigeration.

Tungsten—Tungsten is a ferrite-forming element which will give a controlled transformation steel with 24% δ-ferrite at 17%Cr–4%Ni–3%W. It is an expensive element which does not have any marked effect in improving high-temperature properties.

Vanadium—Vanadium is such a powerful ferrite-forming element that controlled-transformation compositions have high δ-ferrite contents.

Copper—Copper is an austenite-forming element

TABLE VI

Properties of 0·092%C–2·16%Mn–0·18%Si–4·26%Ni–16·92%Cr–1·04%Mo steel

Heat-treatment	Tensile Strength, tons/in²	Proof Stress, tons/in² for extension, %				Limit of Proportionality, tons/in²	Elongation on $4\sqrt{A}$, %	Reduction in Area, %
		0·20	0·10	0·05	0·02			
1000° C A.C.	89·6	15·9	15·4	14·7	13·2	6·4	20·0	40·0
1000° C 1 h, − 78° C 1 h, 400° C 2 h	86·4	52·6	51·2	49·4	46·3	33·6	18·2	51·2
950° C 1 h, − 78° C 1 h, 400° C 2 h	85·4	66·8	62·3	58·5	53·0	42·0	18·0	48·2
1000° C 1 h, 700° C 2 h, 450° C 1 h	71·8	58·4	57·6	56·5	54·6	46·4	18·0	49·2
1000° C 1 h, 700° C 2 h, 450° C 4 h	71·7	60·8	59·6	58·5	56·9	52·8	20·0	51·2

which will produce a ferrite-free controlled-transformation steel at 17%Cr–4%Ni–2½%Cu. The principal disadvantage of this steel is the difficulty of hot working. The main advantage of Cu is its ability to promote considerable secondary hardening, although over-ageing does occur rapidly at 450° C.

Cobalt—Cobalt is an austenite-forming element with a very small effect in depressing the M_s temperature. Although it is expensive it does produce very marked secondary hardening which does not over-age at temperatures up to 500° C. It should be extremely useful as a control element, because it can be used for ferrite control without affecting the M_s very much.

Aluminium—Aluminium is an extremely powerful ferrite-forming element and consequently, cannot be used to produce simple controlled-transformation steels. However, it does produce very marked secondary hardening, and therefore, Al should be a useful alloy addition in a properly balanced composition for producing very high-strength steels.

Development of high-strength steels

Many of the simple compositions indicated above have already found commercial application. Examples are:

(i) Armco 17–4 PH:
  (*a*) 0·04%C–16·5%Cr–3·5%Ni–3·5%Cu[5]
  (*b*) 0·1%C–17%Cr–4%Ni–2%Cu[6]

Whilst (*b*) is a simple transformable steel it also shows ageing.

(ii) Allegheny-Ludlum AM.350:
  0·1%C–17%Cr–4·2%Ni–2·75%Mo[7]
(iii) United Steel: 0·1%C–17%Cr–4%Ni–3%Mn
(iv) United States Steel Corporation:
  16%Cr–6%Ni–1%V

Several more complicated steels have also been described which combine transformation with ageing.

(v) United States Steel Corporation Stainless W:
  0·07%C–17%Cr–7%Ni–0·7%Ti–0·2%Al[8]
(vi) Armco 17–7 PH:
  0·07%C–17%Cr–7%Ni–1·1%Al[8]
(vii) Armco 15–7 PH Mo:
  0·07%C–15%Cr–7%Ni–2·5%Mo 1·2%Al[10]
(viii) Firth Vickers FV.520:
  0·07%Cr max.–1%Si max.–2%Mn max.–14–18% Cr–4–7%Ni–1–3%Mo–0·5%Ti max.[11]

It will be seen that these combine various elements which are effective in producing controlled-transformation and ageing steels. From the discussion that has been given in previous sections of this paper it is possible to understand the changes which occur during heat-treatment and the reason for the mechanical properties obtained. The most recently developed steels show quite appreciably improved mechanical properties and the authors have been interested in investigating the potentialities of the various alloy systems. A convenient starting point was the simple controlled-transformation steel and the two examples mentioned previously were used, namely 17%Cr–4%Ni–3%Mn and 17%Cr–4%Ni–2%Mo. The details of this work provide an interesting demonstration of the effects of alloying elements and indicate the mechanical properties which can be obtained.

Development from 17%Cr–4%Ni–3%Mn

The main disadvantage of this composition is the difficulty of producing satisfactory response to primary tempering because of the lack of ferrite. This can be overcome by substituting some Mo for Mn, which will also improve the tempering resistance. It was shown that, whereas there was quite a marked improvement when 1%Mo is added to the base, a further 1% produces little change. A 0·1%C–17%Cr–4%Ni–2%Mn–1%Mo steel was therefore made which

TABLE VII

Properties of 0·14%C–0·24%Si–15·69%Cr–4·14%Ni–3·0%Mn steel

Heat-treatment	Tensile Strength, tons/in²	Proof Stress, tons/in² for extension, %				Limit of Proportionality, tons/in²	Elongation on $4\sqrt{A}$, %	Reduction in Area, %
		0·20	0·10	0·05	0·02			
1050° C 1 h, 700° C 2 h, 450° C 2 h	77·0	44·4	38·4	32·8	26·8	19·2	29·3	41·7
950° C 1 h, 700° C 2 h, 450° C 2 h	83·0	71·8	68·7	64·8	56·0	36·8	15·3	36·0
925° C 1 h, − 78° C 1 h, 400° C 2 h	93·6	78·0	73·3	67·3	55·2	40·0	19·0	40·0

TABLE VIII

Properties of 0·097%C–2·88%Mn–0·34%Si–3·14%Ni–17·70%Cr–2·14%Cu steel

Heat-treatment	Tensile Strength, tons/in²	Proof Stress, tons/in² for extension, %				Limit of Proportionality, tons/in²	Elongation on $4\sqrt{A}$, %	Reduction in Area, %
		0·20	0·10	0·05	0·02			
1050° C 1 h A.C., – 78° C 2 h, 400° C 24 h	82·4	59·6	48·5	38·7	20·5	12·8	4·9	9·6
1000° C 1 h A.C., – 78° C 2 h, 400° C 24 h	89·8	66·9	55·2	45·0	34·7	16·0	11·6	14·4
975° C 1 h, – 78° C 2 h, 400° C 24 h	97·6	81·6	74·0	63·8	40·6	24·0	11·6	23·2
1050° C 1 h, 700° C 2 h, 400° C 1 h	75·5	59·4	53·2	48·0	40·2	26·4	19·4	36·0
1050° C 1 h, 700° C 2 h, 400° C 6 h	75·8	60·6	53·2	44·6	35·0	19·2	ND*	4·8
1050° C 1 h, 700° C 2 h, 400° C 24 h	72·6	62·0	55·0	44·8	35·6	20·0	2·4	4·8
1050° C 1 h A.C., 700° C 2 h, 450° C 1 h	77·6	68·8	63·0	57·2	47·1	25·6	4·8	7·0
1050° C 1 h A.C., 700° C 2 h, 500° C 2 h	70·1	57·6 YP	57·0	51·8	42·6	21·6	19·9	41·6
1050° C 1 h, 700° C 2 h, 550° C 2 h	63·0	49·4	46·0	41·2	33·4	20·0	20·6	43·2

* Test-piece broke outside gauge length

had 12% δ-ferrite in the 1050° C solution-treated condition. This steel responded satisfactorily to both primary tempering and refrigeration, and typical properties are given in Table VI.

Whilst 80 tons/in² is a quite satisfactory tensile strength, higher strengths can always be utilized in aircraft applications. Since the structure produced is martensitic, one way of increasing the tensile strength is to increase the carbon content. If the carbon content is raised to 0·14% it is necessary to reduce the alloy content to maintain the controlled-transformation characteristics. It was determined that a reduction of 1½%Cr would balance this carbon increase, and the steel developed had had the composition 0·14%C–15·5%Cr–4·2%Ni–3%Mn. Hardness results of 470 and 370 DPN were obtained after refrigeration and primary tempering respectively. Typical mechanical properties are given in Table VII and these confirm the satisfactory response to both types of heat-treatment. If these are compared with the results for a 0·09%C steel in Table II it will be seen that an increase of 0·05%C has produced an increase of roughly 10 tons/in² in the tensile strength.

The other way in which tensile strength can be increased is to make use of secondary hardening. The effects of both Cu and Al have been examined. Experimental work showed that 17%Cr–3%Ni–3% Mn–2%Cu (6% δ-ferrite) and 17%Cr–4%Ni–3%Mn–1%Al (12% δ-ferrite) were the optimum compositions. It will be realized that these preferred compositions are selected from a series of experimental steels on the basis of ferrite content and satisfactory response to heat-treatment. The optimum heat-treatment was determined on the basis of hardness tests as previously described. Typical mechanical properties are given in Tables VIII and IX. It will be seen that both steels

TABLE IX

Properties of 0·09%C–2·8%Mn–0·32%Si–17·48%Cr–4·44%Ni–0·9%Al steel

Heat-treatment	Tensile Strength, tons/in²	Proof Stress, tons/in² for extension, %				Limit of Proportionality, tons/in²	Elongation on $4\sqrt{A}$, %	Reduction in Area, %
		0·20	0·10	0·05	0·02			
1050° C 1 h A.C.	65·8	16·8	15·7	14·5	12·0	7·2	30·6	51·2
1050° C 1 h, 700° C 2 h	59·4	40·8	33·4	27·8	21·8	10·4	16·8	52·8
1050° C 1 h, 700° C 2 h, 450° C 2 h	84·2	73·8	69·8	62·6	52·0	38·4	13·0	32·0
1050° C 1 h, 700° C 2 h, 500° C 2 h	91·2	85·3	80·8	74·9	64·0	46·4	ND*	2·0
950° C 1 h A.C.	69·0	20·9	19·6	18·6	17·3	8·0	18·6	47·2
950° C 1 h A.C., – 78° C 2 h	72·5	49·4	43·2	37·6	29·8	14·4	15·3	47·2
950° C 1h, – 78° C 2 h, 450° C 4 h	93·8	81·8	74·0	65·0	56·4	42·4	12·8	18·4

* Fracture outside gauge length

TABLE X

Properties of 0·09%C–0·28%Si–0·5%Mn–17·12%Cr–4·18%Ni–2·02%Mo–2·02%Co steel

Heat-treatment	Tensile Strength, tons/in²	Proof Stress, tons/in² for extension, %				Limit of Proportionality, tons/in²	Elongation on $4\sqrt{A}$, %	Reduction in Area, %
		0·20	0·10	0·05	0·02			
1050° C 1 h A.C.	83·8	13·8	13·4	13·0	11·9	8·0	27·1	36·0
1050° C 1 h A.C., 700° C 2 h	67·0	44·6	36·0	28·2	21·0	12·0	15·6	51·2
1050° C 1 h, 700° C 2 h, 450° C 1 h	72·4	58·0	55·4	53·8	52·0	43·2	17·8	47·2
1050° C 1 h, 700° C 2 h, 450° C 6 h	75·2	62·2	58·9	56·8	54·0	44·8	19·6	45·2
1050° C 1 h, 700° C 2 h, 450° C 24 h	80·3	70·4	67·5	64·9	60·8	44·8	17·8	40·0
1050° C 1 h, 700° C 2 h, 500° C 1 h	71·1	60·6	58·2	56·6	53·6	41·6	18·8	47·2
1000° C 1 h A.C., − 78° C 2 h	87·2	56·6	50·9	44·0	33·9	17·6	15·6	43·2
1000° C 1 h A.C., − 78° C 2 h, 450° C 6 h	86·8	74·7	70·6	66·6	60·4	51·2	22·5	52·8

respond satisfactory to both types of heat-treatment to give mechanical properties which are considerably higher than the non-ageing steel (Table II). In fact, the mechanical properties of the ageing compositions are surprisingly similar to the higher-carbon modification. On the grounds of superior corrosion resistance and more consistent properties (the M_s control need not be so accurate) the ageing steels are preferred.

Development from 17%Cr–4%Ni–2%Mo

Since the base composition contained ferrite and responded satisfactory to heat-treatment, the developments concerned the addition of age-hardening elements. The effects of Cu and Co were examined and the steels finally developed were 17%Cr–4%Ni–2%Mo–2%Cu and 17%Cr–4%Ni–2%Mo–2%Co. The mechanical properties obtained are given in Tables V and X, and it will be seen that satisfactory response was obtained to both types of heat-treatment. The level of mechanical properties was similar to the 3% Mn ageing steels.

Development of the highest strength composition

Three points which emerged from the previous

TABLE XI

Properties of 0·063%C–2·09%Mn–0·32%Si–16·42%Cr–51·4%Ni–1·51%Mo–0·88%Al–2·03%Co steel

Heat-treatment	Tensile Strength, tons/in²	Proof Stress, tons/in² for extension, %				Limit of Proportionality, tons/in²	Elongation on $4\sqrt{A}$, %	Reduction in Area, %
		0·20	0·10	0·05	0·02			
1050° C 1 h A.C.	55·1	16·5	13·4	11·0	8·7	4·8	46·4	62·4
1050° C 1 h, 700° C 2 h A.C.	60·7	39·8	32·8	26·4	20·4	12·8	16·9	47·2
1050° C, 700° C 2 h, 400° C 2 h	75·9	64·8	59·6	54·8	49·2	36·8	18·2	47·2
1050° C, 700° C 2 h, 400° C 4 h	77·8	66·8	61·8	55·7	48·0	35·2	16·9	43·2
1050° C, 700° C 2 h, 450° C 2 h	90·6	83·4	78·4	73·4	64·8	52·8	1·9	4·1
1050° C, 700° C 2 h, 450° C 4 h	93·6	83·4	78·4	71·4	62·4	38·4	2·9	4·8
1050° C, 700° C 2 h, 500° C 2 h	96·6	87·6	82·8	76·4	66·0	33·6	3·6	4·1
1050° C, 700° C 2 h, 550° C 2 h	92·3	86·2	83·0	79·9	73·8	40·0	1·2	4·1
1050° C, 700° C 2 h, 550° C 4 h	90·0	84·4	81·6	77·6	69·2	41·6	7·9	9·6
1050° C, 700° C 2 h, 575° C ¼ h	88·0	81·1	77·4	72·6	62·8	38·4	15·3	40·0
1050° C, 700° C 2 h, 600° C ¼ h	76·1	69·4	66·7	61·9	53·0	33·6	18·2	36·0
1050° C, − 78° C 2 h, 450° C 4 h	59·1	21·4	17·2	14·2	10·9	7·2	40·0	43·2
950° C, − 78° C 2 h, 450° C 4 h	98·1	80·6	71·2	61·6	50·4	35·2	21·0	32·0

TABLE XII

Properties of 0·07%C–0·34%Si–17·55%Cr–3·48%Ni–2·04%Mn–2·0%Mo–1·21%Cu–2·03%Co steel

Heat-treatment	Tensile Strength, tons/in²	Proof stress, tons/in² for extension, %				Limit of Proportionality, tons/in²	Reduction in Area, %
		0·20	0·10	0·05	0·02		
1050° C 1 h A.C.	65·9	19·2	16·8	14·8	12·1	7·2	51·8
1050° C 1 h, 700° C 2 h	62·6	39·7	32·6	27·0	21·0	10·4	51·8
1050° C 1 h, 700° C 2 h, 450° C 1 h	78·4	67·8	64·2	60·8	56·6	46·4	40·0
1050° C 1 h, 700° C 2 h, 450° C 4 h	81·8	71·7	67·8	64·4	59·8	46·4	14·4
1050° C 1 h, 700° C 2 h, 500° C 1 h	73·0	65·8	62·6	59·9	56·6	49·6	36·0
1050° C 1 h, 700° C 2 h, 550° C 1 h	64·2	59·1	58·6	57·7	55·6	47·2	45·2
1050° C A.C., − 78° C 2 h, 450° C 4 h	64·8	27·0	22·3	18·0	14·4	8·0	45·2
950° C A.C., − 78° C 2 h, 450° C 4 h	88·2	80·8	75·8	69·2	57·6	38·4	36·0

work were that ageing compositions were better than high carbon compositions for high strengths, that Mo gave improved high-temperature properties and that Cu, Al and Co produced age-hardening reactions. It did appear that improved resistance to over-ageing was obtained with two ageing reactions combined in one steel; consequently, Cu–Co and Al–Co steels were examined. The other main feature of these steels was that a low carbon content (0·07% max.) would give improved resistance to weld decay.

In the Al–Co steel the combination of Mo, Al, and low carbon gave a strong ferrite-forming tendency when compared with a 17%Cr–4%Ni–3%Mn base, so a modified base of 16%Cr–5%Ni–2%Mn was adopted. The final composition analysed 0·06%C–16·4%Cr–5·14%Ni–2·09%Mn–1·51%Mo–0·88%Al–2·03%Co. This steel had 14% of δ-ferrite and also had a correctly controlled martensite transformation range. Typical mechanical properties are given in Table XI. It will be seen that very high tensile properties are obtained in both heat-treated conditions, although the ductilities are low at the peak of the age-hardening reaction. With appropriate heat-treatment, however, extremely attractive mechanical properties can be obtained.

Since both Cu and Co are austenite-forming elements, it is not so difficult to produce low-carbon molybdenum-containing steels and the final composition was 17%Cr–3½%Ni–2%Mn–2%Mo–1·2%Cu–2%Co. This had 15% of δ-ferrite and also had a correctly controlled martensite transformation range. Mechanical properties are given in Table XII and it will be seen that this steel also gives high mechanical properties in both heat-treated conditions.

The final modification which has been examined is to introduce Ti or Nb as stabilizing elements. As these have a strong ferrite-forming tendency (which is accentuated by a strong reaction with carbon) it is necessary to provide further austenite stability by modifying the base composition. Steels have been tested which contain 0·3%Ti and are similar to the above high-strength compositions except that the Cr is slightly lowered and the Ni raised. These have given very satisfactory mechanical properties and indicate that there is no difficulty in adding Ti if this should be necessary for maximum resistance to weld decay.

SUMMARY

In order to meet the requirement for a stainless steel which should be soft to allow fabrication, but capable of being appreciably strengthed by a simple heat-treatment, work has been carried out on controlled-transformation stainless steels. The essential feature of these steels is that the martensite transformation range should be controlled just below RT When the steel is cooled to RT it is austenitic and can be fabricated, but transformation will occur if the steel is cooled to − 70° C or heated to 700° C. The latter treatment is effective because precipitation of alloy carbides takes place, which raises the martensite transformation range.

It has been shown that 17%Cr–4%Ni is a very suitable base composition and many steels have been tested to determine the effect of alloying elements on the structural constitution and the position of the martensite transformation range. Satisfactory controlled-transformation steels can be developed from the 17%Cr–4%Ni base composition by making single additions of Mn, Mo, Co, Si, W, and V. These steels all have 0·05–0·10%C, but because the structure is essentially martensitic, higher strengths can be obtained by raising the carbon content, although this will be at some sacrifice in the corrosion resistance, ductility, and welding characteristics.

It has been found that certain alloying elements promote a strong age-hardening reaction and Cu, Al, Co, Ti, and Nb have been studied. The ageing reaction leads to a considerable increase in the strength of the martensitic structure and it is found that when one of these compositions is tested, the final properties are far less dependent on a complete transformation to martensite. Consequently, a controlled-transformation ageing steel would appear to offer the most promise for commercial development.

The essential requirements for one of these controlled-transformation ageing steels are:

(i) A low carbon content
(ii) Sufficient Cr for adequate corrosion resistance
(iii) Nickel to obtain an essentially austenitic structure with a controlled transformation
(iv) Manganese to provide control over transformation temperature without affecting the microstructure
(v) Molybdenum to provide tempering resistance and improved high-temperature strength
(vi) Copper, Co, or Al to provide an ageing reaction; improved properties are obtained when two of these reactions are combined
(vii) Titanium or Nb for maximum resistance to weld decay.

Compositions have been given which meet these requirements together with typical mechanical properties.

ACKNOWLEDGMENTS

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Andrew Carnegie Research Report

The stress-dependent damping capacity of ferromagnetic metals

By G. Sumner, Ph.D., and K. M. Entwistle, Ph.D., M.Sc.

SYNOPSIS

Stress-dependent damping arises in ferromagnetic metals from the irreversible movement of magnetic domain boundaries. This damping would be expected to reach a maximum value at a vibrational stress which is some average measure of the internal stress in the specimen, since the boundary movements are controlled primarily by the internal stress field through which they move. Measurements in torsional vibration at about 20 c/s on pure iron, steel, and nickel reveal a peak in the relation between magnetic damping and vibrational stress which is evident at higher stresses for smaller values of the magnetic permeability, that is, for higher degrees of internal stress. Torsional plastic strain also displaces the peak to higher strains.

A simple analysis using domain theory predicts a relation between stress for peak damping and magnetizing field for maximum susceptibility which is found experimentally to be valid. A search for a gradual migration of the magnetic damping peak to higher strains during fatigue stressing, reflecting dynamic work hardening, leads to the observation of the expected effect in mild steel but not in pure iron.

Some observations are reported on the damping of non-magnetic origin which increases during sustained vibration at high strain and recovers on resting at room temperature. 1559

CONTRIBUTIONS to the stress-dependent damping of metals arise from the irreversible movement of structural discontinuities, particularly dislocations and magnetic domain boundaries. Damping measurements over a range of stress may therefore, in principle, yield information about these structural features. In the present paper the magnetic domains are singled out for study.

Magnetic hysteresis arises from the irreversible movement of domain boundaries by Barkhausen jumps, which cause magnetic remanence. The magnetostrictive strain associated with this residual magnetism represents a permanent change of dimensions following the application and removal of a

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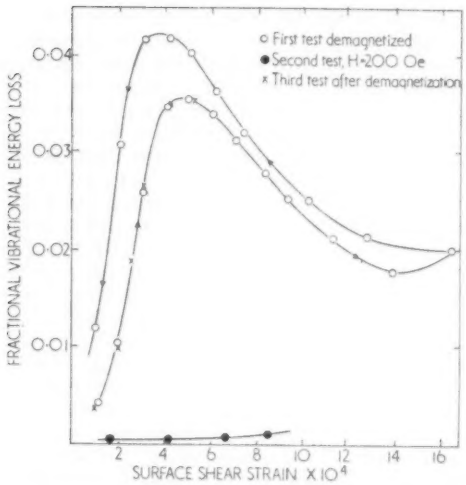


Fig. 1—Relation between damping and surface shear strain for annealed mild steel; frequency 19·1 c/s

magnetizing field. Since stress moves the domain boundaries in a manner broadly similar to that of a magnetic field, a residual strain of the same kind can be produced by the application and removal of stress; thus, except in a material of zero magnetostriction, magnetic hysteresis always connotes elastic hysteresis. This elastic hysteresis would be expected to reflect the internal stress pattern in the metal since the Barkhausen jumps are controlled primarily by the internal stress field through which the domain boundaries move.

The object of the work to be described was to measure the damping capacity of ferromagnetic specimens in torsional vibration over a wide range of strain to explore the extent to which the magnetic hysteresis damping at room temperature reflects the state of internal stress in iron and steel, and in particular that developed by fatigue stressing.

DAMPING MEASUREMENT

A damping technique was developed specifically for this investigation and is fully described in a paper[1] published elsewhere. The method is based on a vibrating system, developed in an earlier investigation,[2] comprising a cylindrical specimen 6 in. long × 0·2 in. dia. with enlarged tapered ends. One end is firmly fixed to a massive concrete bed and the other is fitted with an inertia which defines the torsional vibration frequency to be about 20 c/s. The specimen is maintained in vibration by a moving coil exciter fixed to the oscillating inertia. A crystal, also coupled to the inertia, generates a voltage at the specimen's vibration frequency; this is amplified, adjusted in phase and supplied to a power amplifier, which feeds current to the exciter exactly in phase with the angular velocity of the inertia. In this condition the energy dissipated by the specimen per cycle, expressed as a fraction of the maximum energy, is given by:

$$\frac{\Delta E}{E} = \frac{\pi K i_m}{\frac{1}{2} I \omega^2 \theta_m},$$

where K is the force constant of the exciter, i_m the maximum exciter current, I the effective moment of inertia of the vibrating system, ω the angular vibration frequency, and θ_m the maximum angular displacement of the inertia. All these quantities are measurable. With the apparatus in its present form it is possible to supply a maximum power of only 4 W to the exciter, of which only about 1 W is dissipated by the specimen. This gives a maximum shear strain of about 2×10^{-3} (10 tons/in²) in mild steel but modifications are in hand to increase this figure.

Variation of damping with vibrational strain for mild steel

The fully developed apparatus was used to measure the variation of damping capacity with shear strain for mild steel. The specimen* was prepared with a fine-ground finish and annealed *in vacuo* at 625° C for 18 h to relieve machining stresses. The specimen was surrounded in the damping apparatus by a water-cooled solenoid which served either to de-magnetize the specimen or to magnetize it to any desired intensity.

The important characteristics which are shown by the damping curves of Fig. 1 are:

(i) There is a pronounced peak in the damping-strain curve which is completely removed by an axial magnetic field of about 200 Oe; this establishes that the peak is of magnetic origin and confirms the observations of Boulanger[3]

(ii) The damping of an initially demagnetized specimen measured with rising amplitude of vibration is lower than that recorded with a sequence of falling amplitudes. In the region of the peak damping, about 2 000 vibration cycles must elapse before a stable damping is attained at a particular vibration amplitude. If, for example, the applied torque is reduced after vibration has been maintained for a long period, the damping is observed to fall over a period of about 2 min at 20 c/s but thereafter negligible damping change is detected over 10^6 cycles

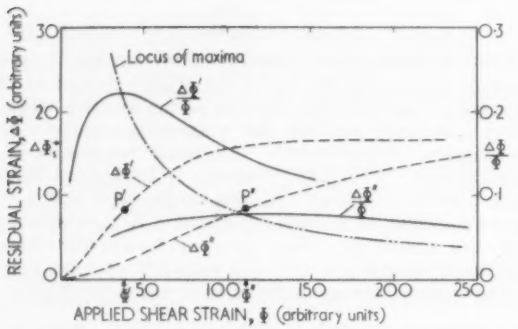
(iii) Demagnetization following vibration causes the damping curve to recover its initial form; this shows that, in this specimen, no permanent change of specimen damping is produced by about 10 000 cycles of vibration at a surface shear strain of 16×10^{-4}, although a recoverable change of magnetic condition arises.

Observations (ii) and (iii) show that the form of the damping curve is sensitive to the initial magnetic condition of the specimen; because of this, in all the tests which are reported the specimen was demagnetized before each damping-strain curve was measured.

Analysis of magneto-mechanical damping peak

The relation between magnetic damping and vibrational shear strain could be calculated precisely only if the internal stress and domain patterns of the specimen were known; however, the general form of the curve can be explained on the following lines.

*BISRA Catalogue of Irons and Steels for Research Purposes. BKK(En4) percentage analysis: C 0·28, Si 0·2, Mn 0·79, S 0·036, P 0·030, Cu 0·12, Ni 0·14, Cr 0·10, N 0·003.



Superscripts " refer to specimen with greater internal stress than tha marked '

Fig. 2—Relation between residual torsional strain $\Delta\Phi$ and the prior applied torsional strain Φ for ferromagnetic wires. Full curves are the derived relation between $\Delta\Phi\Phi$ and Φ

Magnetization, i.e. magnetic domain alignment, can be secured by the application of either stress or a magnetic field H; at low magnetizations the effects of H and stress on the magnetic intensity are proportional. The application and removal of H leaves remanent magnetism which has a residual strain associated with it because of magnetostriction. Similarly a half-cycle of stress will give rise to a residual strain of the same kind which contributes to the width of the mechanical hysteresis loop. Whatever the magnetostrictive characteristics of the metal, the residual strain will always have the same sense as that of the applied stress.

Considering now the case of torsion, twisting to a shear strain ϕ produces helical magnetization along a principal stress direction. Remanent magnetism in this direction after releasing the applied torque causes a residual twist $\Delta\phi$ through the Wiedmann effect. The variation of $\Delta\phi$ with ϕ will be similar to the relation between remanent intensity of magnetization J_r and applied field H because $\Delta\phi$ is related to J_r and ϕ is related to H. The J_r–H curve is S-shaped[4] with zero slope at the origin and the measurements of Becker and Kornetzki[5] show that the $\Delta\phi$–ϕ curve for iron and nickel wires has the same general form, shown diagrammatically in Fig. 2.

$\Delta\phi$ is the width of the magneto-mechanical hysteresis loop and if it is assumed that the shape of the loop is independent of ϕ then the area

$$\Delta E = kG\phi\Delta\phi$$

where G is the shear modulus and k depends on the shape of the loop and has a maximum value of 2 for a parallelogram. The maximum strain energy, E is about $\frac{1}{2}G\phi^2$ per unit volume so that

$$P_m = \frac{\Delta E}{E} = 2k\frac{\Delta\phi}{\phi},$$

where P_m is the magnetic damping and $2k$ is probably slightly greater than 2 but cannot exceed 4. We expect, then, that the variation of P_m with ϕ is similar to the variation of $\Delta\phi/\phi$ with ϕ. The S-shape of the $\Delta\phi$–ϕ curve therefore means that P_m goes through a maximum at the value of ϕ at which $\Delta\phi/\phi$ is a maximum, that is where the line from the origin is a tangent to the $\Delta\phi$–ϕ curve, for example at P' on curve $\Delta\phi'$ in Fig. 2. At high strains $\Delta\phi$ is constant so P_m varies as $1/\phi$ provided that k does not change. This analysis shows that the width of the torsional magneto-mechanical hysteresis loop cannot exceed the Wiedmann strain at saturation, which is about 12×10^{-6} for iron.[5] Normally the greatest width will be less than this because the remanent intensity of magnetization is less than the saturation value.

Ninety-degree domain boundaries seek a position of zero stress, so, in a material of positive magnetostriction, an applied tensile stress equal in magnitude to the height of a compressive internal stress peak will move the boundary to the point of maximum compressive stress and a further slight increase will cause an unstable Barkhausen jump to a point where the resultant stress is again zero. Hysteresis arises because the reverse jump on unloading takes place at a lower stress. Magnetic damping therefore has its origin in the stress-induced movement of 90° boundaries over asymmetrical internal stress peaks. During the first few vibration cycles of a specimen which initially is demagnetized the domain pattern will be modified by the alternating stress but ultimately, during sustained vibration at constant strain, each boundary will make repeated forward and reverse excursions over a particular stress peak and in this way the specimen can dissipate energy indefinitely at a constant rate. For mild steel it appears that about 2 000 cycles are required for this state to be attained; further, the observation that the same damping curve is always observed after demagnetization following vibration must imply that the vibration process at strains in the region of the peak does not modify the internal stress pattern.

We have seen that if J_r is the remanent intensity of magnetization produced by a field H, J_r/H is closely related to the magnetic damping P_m. Now there is evidence[4] that J_r/H reaches a maximum at the same value of H as does the magnetic susceptibility $K = J/H$. This leads to the conclusion that the value of $H(H_m)$ which gives maximum susceptibility causes the same degree of domain boundary movement as the applied stress $G\phi_m$ for peak damping. The relation between H_m and $G\phi_m$ can be derived by comparing the magnetic field and the stress required to move a 90° boundary the same distance. If the internal stress is not too high, so that the magnetoelastic energy is significantly less than the crystal anisotropy energy, iron domains are magnetized in a [100] direction. The field H required to bring a boundary to a compressive internal stress peak of height σ, where both H and σ act normally on the boundary, is

$$HJ_s = \frac{3}{2}\lambda_{100}\sigma$$

where J_s is the saturation intensity and λ_{100} is the saturation magnetostriction in the [100] direction. Alternatively an applied tensile stress σ will reduce the resultant stress to zero and effect the same boundary movement. Hence H and σ produce the same effect if

$$H/\sigma = \frac{3}{2}\frac{\lambda_{100}}{J_s}.$$

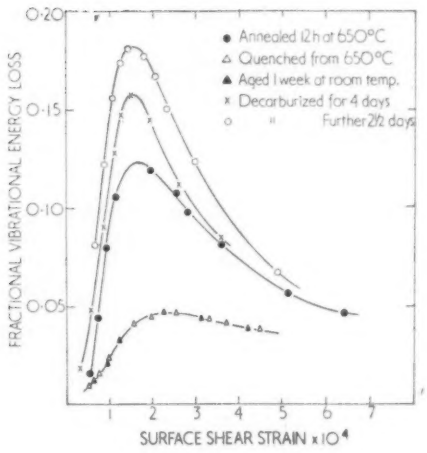

FRACTIONAL VIBRATIONAL ENERGY LOSS
SURFACE SHEAR STRAIN x 10⁴
● Annealed 12 h at 650°C
△ Quenched from 650°C
▲ Aged 1 week at room temp.
× Decarburized for 4 days
○ " Further 2½ days

Fig. 3—Relation between damping and surface shear strain for a deoxidized high-purity iron specimen given treatments indicated, frequency about 18·7 c/s

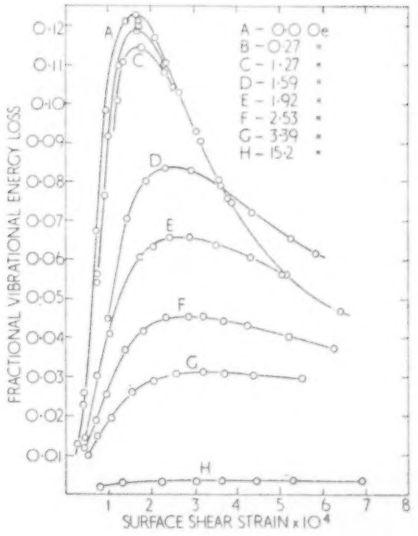


Fig. 4—Effect of a progressively increased axial magnetizing field on damping-surface shear strain curve for deoxidized high-purity iron

In the case of a pure shear stress τ, the tensile stress component is τ and if this opposes the internal compressive stress,

$$\frac{H}{\tau} = \frac{3}{2}\frac{\lambda_{100}}{J_s}.$$

For polycrystalline specimens this simple analysis will be complicated by variation of the angle between the applied effects (H and τ) and the [100] direction from grain to grain in that some averaging factor must be inserted. In the absence of preferred orientation this will be the same for H and τ and hence will not affect their ratio. At high internal stress levels the expression for H/τ will be invalid since the magnetization vectors may not lie in a [100] direction and crystal anisotropy effects must be considered. This is necessary if $\frac{3}{2}\lambda_{100}\sigma$ approaches K_1, the anisotropy constant; that is at a stress of about 200 000 lb/in² for iron, which is well in excess of the highest values of stress for peak damping encountered in the work now described.

Taking J_s for iron to be 1 717 cgs units and the admittedly approximate figure of

$$\lambda_{100} = 18 \times 10^{-6},$$

then

$$\frac{H}{\tau} = 1{\cdot}57 \times 10^{-8}\ \text{Oe dyn}^{-1}\,\text{cm}^2.$$

If this relation is valid then the value of H at maximum magnetic susceptibility should equal $1{\cdot}57 \times 10^{-8}\tau$, where τ is the vibrational shear stress (dyn/cm²) for peak magnetic damping. The comparison must be based on the damping P_h for homogeneous stress. This can be derived from the measured damping P for a solid cylinder by the relation:

$$P_h = P + \frac{\phi}{4}\cdot\frac{dP}{d\phi}.$$

An uncertain error is involved here in assuming that there is no radial interaction between domain movements in adjacent cylindrical elements.

An increase of internal strain will raise the compressive internal stress peaks and hence a higher applied stress $G\phi$ will be required to generate a given value of $\Delta\phi$, that is the $\Delta\phi$–ϕ curve will be sheared to higher strains, for example from curve $\Delta\phi'$ to $\Delta\phi''$ in Fig. 2. The strain for peak damping ϕ'' is thereby increased and the magnitude of the peak damping, proportional to the gradient of OP'', is reduced. If the sole effect were shearing of the curve, the locus of the $\Delta\phi/\phi$ maxima would be a rectangular hyperbola, chain-dotted in Fig. 2, and the peak magnetic damping would vary similarly. In practice, secondary variations of the limiting value of $\Delta\phi$ and possibly the shape of the $\Delta\phi$–ϕ curve will cause the locus to diverge from this law. In general, however, the expected effect of an increase of internal strain will be to move the magnetic damping peak to a higher strain and to reduce its height.

Experiments which were carried out to check the predictions developed here are described in the next section.

Damping characteristics of specimens of different magnetic characteristics

To test the validity of the foregoing simplified analysis, the damping measurements on mild steel were repeated on a magnetically softer material, pure iron, and normalized 0·42% carbon steel, which is magnetically harder. In addition, measurements were made on nickel to show that the behaviour of mild steel is characteristic of ferromagnetic metals generally.

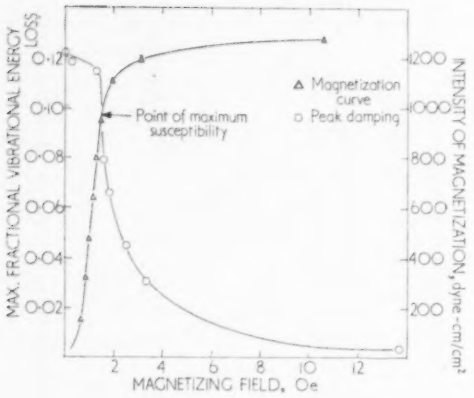


Fig. 5—Comparison between variation of peak damping with magnetizing field and magnetization curve for deoxidized high-purity iron. Steepest fall of damping coincides with the point of maximum susceptibility

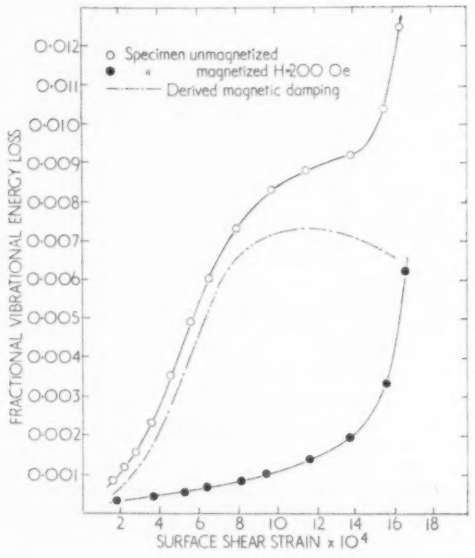


Fig. 6—Damping characteristics of an 0·42% carbon steel specimen in the magnetized and unmagnetized conditions; frequency 19·2 c/s

Pure iron

The damping variation with vibrational strain for a deoxidized high-purity iron* specimen is shown in Fig. 3 for the following conditions; annealed *in vacuo* at 650° C for 12 h; quenched from 650° C; aged one week at room temperature and decarburized in wet hydrogen for the times indicated at 760° C. A damping peak of magnetic origin is evident in all the curves and rises to a greater height and occurs at a lower strain than that measured for annealed mild steel. This supports the prediction that lower internal stress gives a higher peak and displaces it to a lower strain.

Although the decarburization treatment would not completely eliminate interstitial solutes it is reasonable to assume that it would produce the softest magnetic condition of those referred to in Fig. 3; this is consistent with its giving the highest damping. The lowest damping is possessed by the quenched specimen which will be the most highly stressed. Quench ageing produces no detectable damping change.

Becker and Kornetzki[5] measured a maximum value of 0·08 for $\Delta\phi/\phi$ in carbonyl iron at $\phi = 0{\cdot}7 \times 10^{-4}$. On the approximate analysis set out earlier this is equivalent to a peak damping of about 0·2 which is of the same order as the peak damping of 0·18 for the decarburized pure iron specimen at $\phi = 1{\cdot}4 \times 10^{-4}$. It is not unreasonable to expect Becker and Kornetzki's wires to be the softer magnetically, which would explain the lower strain at the peak in their case.

Figure 4 shows the effect on the damping curve for pure iron of axial magnetic fields H, which are seen to reduce the peak damping and displace it to higher strains. This is explained by an increase with H of the number of Barkhausen jumps produced by the field, which then holds the domain boundaries in position and renders stress-induced movement progressively less frequent. This causes a reduction of magnetic damping ultimately to zero at axial saturation. The relation between magnetic intensity and damping is well illustrated in Fig. 5 in which two curves are shown, relating peak damping (P_m) and H, and intensity of magnetization (J), and H. A steep fall of P_m occurs at a value of H which coincides closely with that giving maximum susceptibility, i.e. the point where H produces a large number of Barkhausen jumps. Allowance was made in the magnetic measurements for the demagnetization effect of the open magnetic circuit.

0·42% Carbon steel

The relation between damping and vibrational strain for an initially demagnetized specimen of normalized 0·42% carbon steel* is given in Fig. 6 together with the curve for the magnetically saturated condition. A peak of magnetic origin is evident at a higher strain and of lower height than that for mild steel, reflecting the higher internal stress.

Nickel

The magnetic damping of nickel in the as-machined state and after annealing *in vacuo* at about 950° C was obtained by subtracting the ordinates of the curves measured in the magnetized and unmagnetized states. The difference between the curves showed a

* BISRA Catalogue of Irons and Steels for Research Purposes, 1956 AHK 9. Percentage analysis: C 0·0028, N 0·0014.

* Made to specification of percentage analysis: C 0·42, Si 0·32, Mn 0·99, S 0·032, P 0·031, Ni 0·09, Cr 0·06.

TABLE I

Peak heights and position of magnetic damping peaks for three ferrous specimens

Material	*Peak height*	*Surface shear strain at peak*
Pure iron	**0·18**	**1·4 × 10⁻⁴**
Mild steel	**0·035**	**4·6 × 10⁻⁴**
0·42%C steel	**0·0073**	**11·5 × 10⁻⁴**

peak in both cases, supporting the contention that this is a general characteristic of ferromagnetic materials. The stresses for peak magnetic damping and the peak heights are given in Table II. A very large increase of damping is brought about by annealing the worked specimen, giving a maximum damping of 0·52 at a surface shear strain of about 4×10^{-5}. In this case the damping falls to less than $\frac{1}{50}$th of this in the magnetized state. Becker and Kornetzki found a maximum value of $\Delta\phi/\phi = 0{\cdot}21$ for nickel wire at $\phi = 4 \times 10^{-5}$. The analysis set out earlier predicts a damping of about 0·5 corresponding to this $\Delta\phi/\phi$ value if $2k$ is a little more than 2; this is in excellent agreement with the present results.

An interesting feature of the behaviour of the annealed nickel specimen in the damping machine is the instability which sets in at a surface strain of 12×10^{-5}. At this point the amplitude of vibration begins to rise in an uncontrolled manner to about 45×10^{-5}. In this range of strain the damping is falling more steeply than $1/\phi$ so the alternating torque required to maintain the vibration falls with increasing amplitude. Because the feedback circuit preserves constant applied torque, the strain rises until the development of the non-magnetic damping causes the damping torque to rise again to a value equal to the applied torque. The onset of instability must correspond to a change of shape of the mechanical hysteresis loop of such a nature that a large increase of strain causes only a small increase of area of the loop; the loops reported by Becker and Kornetzki[5] show that this arises by the development of spikes enclosing negligible area. The instability would be expected to set in just below the point at which stress has just completely aligned the domains; Becker and Kornetzki's $\Delta\phi$–ϕ curve for nickel shows that this is effected at about $\phi = 12 \times 10^{-5}$, in good agreement with the observed strain at which the uncontrolled rise of vibration amplitude begins.

Relation between magnetic properties and damping characteristics

The height and position of the magnetic damping peaks for the three ferrous specimens, increasing in degree of internal stress in the sequence pure iron–mild steel–0·42% carbon steel are shown in Table I, which confirm the prediction that the magnetic peak is lower and occurs at higher strains in more highly stressed materials.

TABLE II

Height and position of magnetic damping peak for nickel

Condition	*Peak height*	*Surface shear strain at peak*
As-machined	**0·031**	**36 × 10⁻⁴**
Annealed at 950° C	**0·52**	**4·0 × 10⁻⁴**

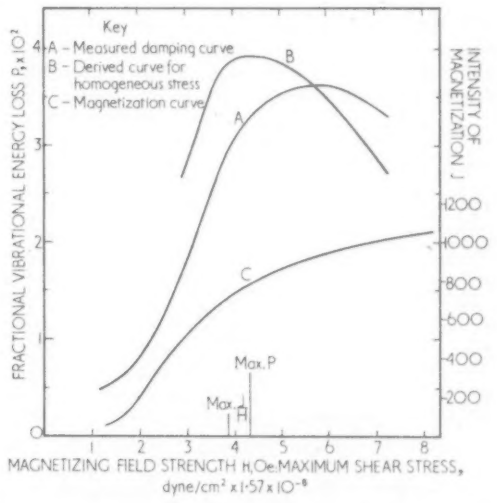

Key
A – Measured damping curve
B – Derived curve for homogeneous stress
C – Magnetization curve
FRACTIONAL VIBRATIONAL ENERGY LOSS P, x 10²
INTENSITY OF MAGNETIZATION J
Max.P
Max. H
MAGNETIZING FIELD STRENGTH H, Oe. MAXIMUM SHEAR STRESS, dyne/cm² x 1·57 x 10⁻⁸

Fig. 7—Relation between damping and magnetization curves for mild steel. Damping is plotted against shear stress in dyn/cm² multiplied by $1{\cdot}57 \times 10^{-8}$

Corresponding figures for as-machined and annealed nickel in Table II give further support to this conclusion and also show the high sensitivity of this material to slight plastic strain.

The magnetic damping-stress curve for annealed mild steel together with the derived curve for homogeneous stress are shown in Fig. 7 plotted in terms of $1{\cdot}57 \times 10^{-8}\tau$, where τ is the vibrational stress in dyn/cm. Superimposed on the graph is the J–H curve for the same specimen. With this choice of abscissae the point of maximum damping should coincide with the point of maximum susceptibility. The agreement is seen to be very good.

Similar comparisons were made for pure iron, 0·42% carbon steel, and annealed nickel. The results are collected in Table III. The comparison in the case of nickel was based on

$$\frac{H}{\tau} = \frac{3}{2} \cdot \frac{\lambda_{111}}{J_s} = \frac{3}{2} \cdot \frac{26 \times 10^{-6}}{484} = 8{\cdot}07 \times 10^{-8} \text{ Oersted dyn}^{-1}\text{ cm}^2.$$

Considering the approximate nature of the assumptions made, the agreement is remarkably good and gives strong support to the suggested domain mechanism for the magnetic damping.

Both stress and H will also move 180° boundaries, but this will not contribute to the over-all strain and therefore cannot give rise to elastic hysteresis. However, 180° reversals give a large contribution to J,

TABLE III

Comparison between measured value of H for maximum susceptibility, and value predicted from damping curves

Material	*Homogeneous strain for peak damping*	*τ, dyn/cm²*	*Predicted H for maximum K, Oe*	*Measured value of H, Oe*
Pure iron	**11 × 10⁻⁴**	**8·17 × 10⁷**	**1·3**	**1·5**
Mild steel	**3·55 × 10⁻⁴**	**27·7 × 10⁷**	**4·3**	**3·9**
0·42%C steel	**8·4 × 10⁻⁴**	**65·6 × 10⁷**	**10·3**	**11·0**
Annealed nickel	**3 × 10⁻⁴**	**1·77 × 10⁷**	**1·4**	**2·1**

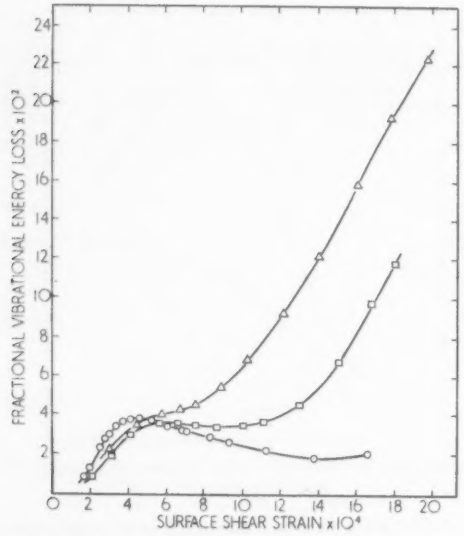


Fig. 8—Relation between damping and surface shear strain for a mild-steel specimen

and therefore the agreement in Table III must mean that 90° jumps predominate. It would be of interest to compare damping with magneto-resistivity, which is also unaffected by 180° reversals.

Effects of sustained vibration at high strain on the damping characteristics

The established dependence of the position and height of the magnetic damping peak on internal stress led to a search for a displacement of the peak to higher strains by high amplitude of vibration. This would reflect the development of internal stress during fatigue straining.

Figure 8 shows damping curves for mild steel up

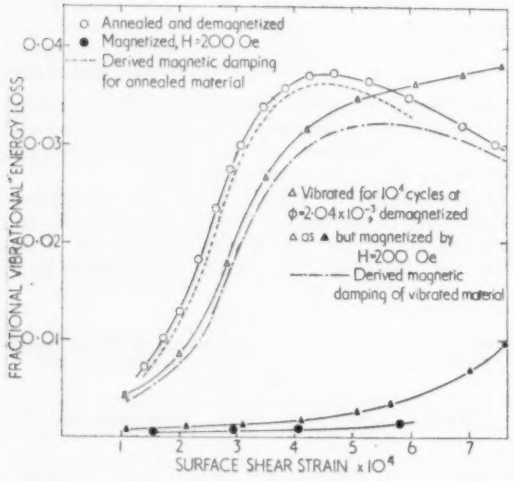

Annealed and demagnetized
Magnetized, H=200 Oe
Derived magnetic damping for annealed material
Vibrated for 10⁴ cycles at φ=2·04 x 10⁻³, demagnetized
as ▲ but magnetized by H=200 Oe
Derived magnetic damping of vibrated material
FRACTIONAL VIBRATIONAL ENERGY LOSS
SURFACE SHEAR STRAIN x 10⁴

Fig. 9—Effect of vibration at high strain on magnetic damping of mild steel

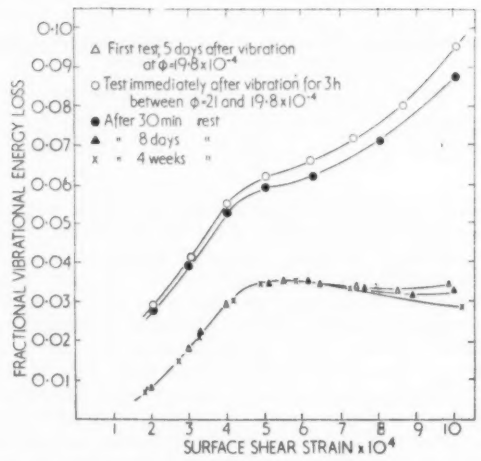

First test, 5 days after vibration at φ=19·8 x 10⁻⁴
Test immediately after vibration for 3h between φ=21 and 19·8 x 10⁻⁴
After 30 min rest
8 days
4 weeks
FRACTIONAL VIBRATIONAL ENERGY LOSS
SURFACE SHEAR STRAIN x 10⁴

Fig. 10—Damping curves showing recovery of vibrationally-induced damping in mild steel

to surface shear strains of about 2×10^{-3}. This is not quite high enough to produce fatigue failure but modifications are in hand to increase the maximum power input to the specimen, at present only about 1 W, to a level which will comfortably fracture the specimen. In the annealed condition a steep rise of damping during vibration begins at a surface strain of about 18×10^{-4}; with the equipment delivering maximum undistorted power to the exciter, a stable damping of 0·29 was reached at a strain of $19 \cdot 3 \times 10^{-4}$. The damping curve measured immediately after vibration was stopped and the specimen had been demagnetized is shown in Fig. 8. An increase of

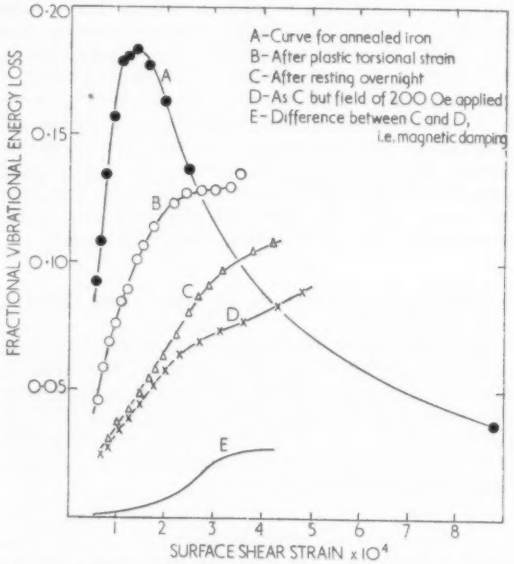

A–Curve for annealed iron
B–After plastic torsional strain
C–After resting overnight
D–As C but field of 200 Oe applied
E–Difference between C and D, i.e. magnetic damping
FRACTIONAL VIBRATIONAL ENERGY LOSS
SURFACE SHEAR STRAIN x 10⁴

Fig. 11—Effect of torsional plastic strain on the damping of deoxidized high-purity iron. Curve *A* is almost wholly magnetic damping in the region of the peak: after straining, magnetic damping is reduced to curve *E*

damping compared with the annealed state is evident involving a factor of about eight, for example, at $\phi = 16 \times 10^{-4}$. Resting for five days at room temperature causes appreciable recovery of damping towards that of the annealed state.

Measurements with the specimen magnetically saturated proved most of the damping induced by vibration to be of non-magnetic origin but Fig. 9 shows that a detectable change of magnetic damping occurs. Vibration for 10 000 cycles at a strain of $20{\cdot}4 \times 10^{-4}$ is seen to reduce the height of the magnetic damping peak for the unvibrated state and displace it from a strain of $4{\cdot}65 \times 10^{-4}$ to $5{\cdot}7 \times 10^{-4}$. This shift is consistent with the development of internal stress although to a lesser degree than was anticipated. It may well be that a bigger effect is observed when strains are attained which will ultimately produce fatigue fracture.

The recovery at room temperature in the damping induced by vibration is shown in Fig. 10. It is evident that the recovery reaches a limit at lower strains earlier than that at higher strains, where the damping is still falling slowly after four weeks. The recovery effect involves only the contribution of non-magnetic origin; its kinetics and possible relation to strain ageing are being investigated.

Vibration of pure iron for 32 000 cycles at a surface strain of $10{\cdot}15 \times 10^{-5}$ produced no change of magnetic damping, although a large increase of non-magnetic damping arose which recovered almost completely to the annealed curve after heating for 4 h at 100° C. On the other hand, torsional plastic strain produced a large reduction of magnetic damping together with an increase of non-magnetic damping which recovered at room temperature (Fig 11).

The general conclusion to be drawn at this stage is that careful measurement of the magnetic damping in mild steel at regular intervals during a fatigue test may yield information of value, but final judgment must be deferred until higher vibrational strains have been achieved.

DISCUSSION

A notable feature of the damping results in the present work is the high order of magnitude of the energy dissipation, both at lower amplitudes, where the magnetic damping predominates, and at higher stresses, where the damping arises presumably from a dislocation mechanism. The damping values for mild steel at high stress are in good agreement with those measured by Lazan[6] and confirm the reliability of his rotating cantilever method for damping studies on steel at fatigue stresses. Lazan's curves do not often show a magnetic peak, although in one case[7] it is clearly present.

An earlier attempt to predict the form of the relation between magnetostrictive damping and stress by Cochardt[8] makes assumptions which differ from those made in the present analysis and will now be discussed. Cochardt assumes that, for a homogeneously stressed specimen, the damping curve can be represented by two distinct regions which meet at a critical stress σ_c, defined as that stress which just aligns all the domains. Below σ_c ΔE is assumed to vary as σ^3 and above σ_c ΔE is constant. A curve which includes σ_c as an arbitrary parameter is calculated on this basis and is compared with the measured curve, using a value for σ_c equal to the measured vibrational stress at peak damping.

The analysis in the present work shows that, referring to Fig. 2, the condition for peak damping, $dP/d\phi = 0$, occurs when $d/d\phi\ (\Delta\phi) = \Delta\phi/\phi$, i.e. where the tangent from the origin touches the $\Delta\phi/\phi$ curve. This point must be below the stress required to produce complete domain alignment, equal to the applied stress which just produces $\Delta\phi_s$, the maximum after-effect In the case of nickel the ratio of stresses is over three; the factor may be less than three in other materials but cannot approach unity as assumed by Cochardt. Further, as the results of Becker and Kornetzki show, two regions are inadequate for a full description of the damping curve. In general ΔE varies as σ^n, where n falls continuously from 3 at low stress to 0 at high stress; thus $n = 2$ at the peak and falls from 2 to 0 over a fairly wide range of stress above it. Cochardt assumes that n is uniformly zero in this region which, as his Fig. 5 demonstrates, will cause his predicted curve to fall more steeply than the measured curve.

The magnetic damping is discussed in the present work on the basis of Becker and Kornetzki's classical after-effect measurements on ferromagnetic wires. Here a continuous variation of n with σ is implied. An important conclusion which emerges from the present work is the wealth of information on magnetostrictive damping which can be inferred from Becker and Kornetzki's results and which have now been confirmed experimentally.

ACKNOWLEDGMENTS

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An electrolytic jet machining technique for the production of thin foils of steel

By P. M. Kelly, B.A., and J. Nutting, M.A., B.Sc., Ph.D.

SYNOPSIS

To obtain thin foils of steel and other hard ferrous alloys, suitable for direct examination by transmission electron microscopy, it is necessary to produce thin sheets (0·05 mm thick) of the metal from relatively thick material, without destroying or modifying the structure of the metal in any way. An electrolytic jet machining apparatus for producing these uniform thin sheets of steel is described and some electron micrographs of thin foils made from these sheets by controlled electropolishing are shown. 1676

IN THE PAST FEW YEARS several workers[1–6] have developed and used techniques for producing metal foils thin enough to be used instead of replicas for the examination of metal structures in the electron microscope. In all these cases the starting point was a sheet of metal 50–200μ thick. The chief disadvantage of these techniques is that they cannot be applied directly to the harder, less ductile metals and alloys, in particular the majority of steels, since it is difficult and in some cases impossible, to produce uniform thin sheets (of the order of 100μ thick) by mechanical working. The apparatus described below was designed and constructed to overcome this difficulty. Using this technique, followed by controlled electropolishing, thin foils of steel can easily be made from bulk specimens of the material without destroying or modifying the structure of the metal in any way.

EXPERIMENTAL PROCEDURE

The stages in the production of thin foils from bulk material using this technique may be summarized as follows:

(i) Preparation of the initial specimen blank
(ii) Electrolytic machining of the blank down to 50–100μ thick
(iii) Electropolishing to clean the surface and give the final reduction in thickness to 20μ
(iv) Preparation of the foils from the thin metal sheet by electropolishing using the technique developed by Brandon and Nutting.[7]

A specimen blank 3 cm square and 4–6 mm thick is cut mechanically from the material and ground to about 1 mm in thickness on a surface grinder. Recent work by Samuels and Wallwork[8] and others indicate that the deformation caused by surface grinding is not likely to extend more than 50–100μ below the surface. The electrolytic machining operation can be made to remove considerably more than this from each side of the specimen so that the final thin sheet produced will in fact be free from the effects of the grinding and cutting operations. If the specimen blank can be heat treated after its preparation, the thickness may be reduced by hot or cold rolling. In either case the specimen must be of uniform thickness as variations in thickness persist throughout the electrolytic machining operation.

The electrolytic jet machining apparatus, a sketch of which is shown in Fig. 1, is similar in principle to that used by Glaysher and Farmer[9] to turn the gauge length of Chevenard tensile specimens, and is built mainly of Perspex and Tufnol, both of which resist the corrosive action of the electrolyte used. Electrolyte consisting of three parts concentrated HCl and one part concentrated aqueous solution of $CuCl_2$, flows through a copper tube-cathode, which is connected to a glass nozzle about 1 mm dia. The specimen holder is mounted perpendicular to the jet assembly, with the specimen about 2 mm from the end of the glass nozzle. The jet traverses backwards and forwards across the specimen at constant velocity 30 times a minute, while the specimen is moved slowly (1 cycle in 6 min) up and down, also at constant velocity. Both these motions are imparted by very accurate cardioid-shaped cams with 1-in. throw, driven at the appropriate speeds by a geared constant-speed motor. Provided the vertical traverse is sufficiently slow in relation to the horizontal traverse, so that each successive stroke of the jet overlaps the preceding one to a certain extent, there is no need for synchronization of the two motions and no risk of the jet cutting a ' zig-zag ' trace across the specimen. A potential of 40–50 V and a current of 2 A is used, giving a rate of metal removal of about 10μ/min over the 1-in. square ' window ' cut in the specimen. Since the rate of metal removal is fairly constant at 300μ/Ah, even for different steels, the thickness of the specimen at any time during the thinning operation can be estimated from the initial thickness, the machining time, and the current.

The uniformity of the final thin sheet depends on the thickness of metal removed and the accuracy of the operation is best expressed as a percentage of the amount dissolved. The average accuracy of the present apparatus is better than 2% of the amount dissolved.

The thinned specimen is removed from the apparatus at a thickness of 50–100μ, is washed and then electropolished in 10 : 1 acetic–perchloric acids to reduce the thickness to 20μ. The final reduction is

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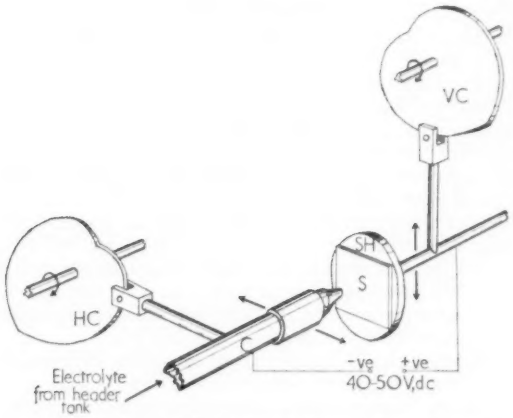

C Cathode SH Specimen holder S Specimen
HC Horizontal traverse cam J Jet VC Vertical traverse cam

Fig. 1—Electrolytic jet machining apparatus

done in this way, as the degree of control of the thinning is better using ordinary electropolishing.

The final thinning technique is the ' figure of eight ' method developed by Brandon and Nutting[7] for producing thin foils in iron. The specimen is blanked off on one side and electropolished in 10 : 1 acetic–perchloric acids, using twin vertical cathodes mounted on either side of the specimen. When perforation has extended sufficiently, as seen through inspection windows in the cathodes, the specimen is removed and washed in methyl alcohol. Small pieces of foil are cut off under methyl alcohol and mounted on electron microscope grids. The foils must be examined soon after preparation or stored in a vacuum desiccator to prevent the formation of surface oxide films.

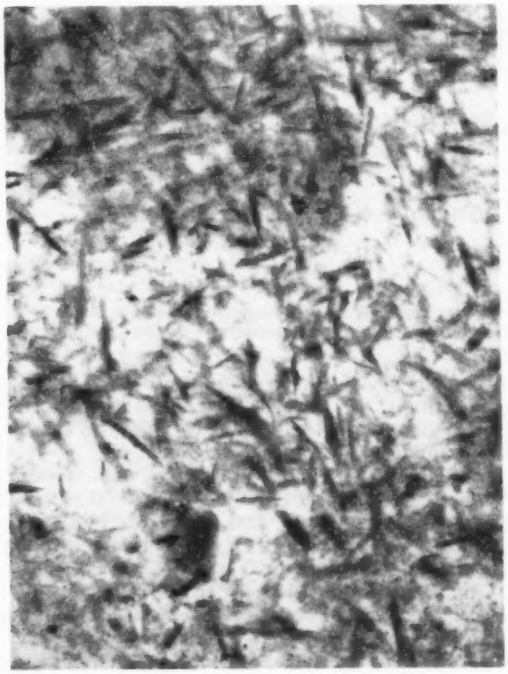

Fig. 2—Thin foil of a 0·14%C steel quenched and tempered for 1 h at 100° C, showing carbide precipitates × 80 000

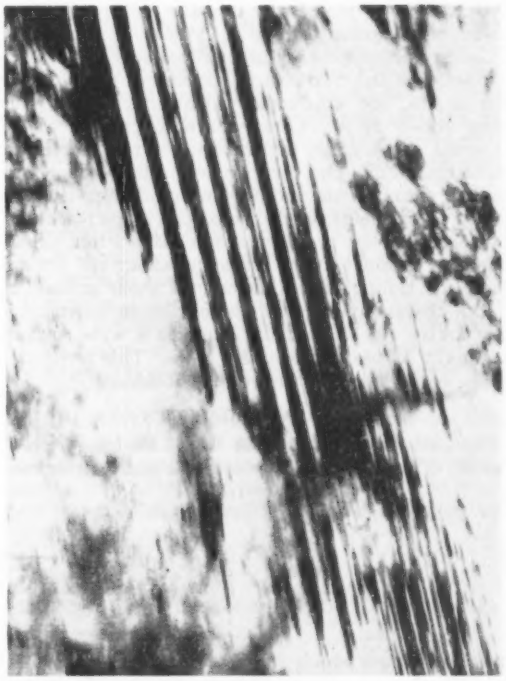

Fig. 3—Thin foil of a quenched 0·98%C steel showing striations running across a single martensite needle × 100 000

Thin foils of 0·14%C and 0·98%C steels in various conditions of heat treatment have been prepared from bulk specimens using the method described above. These have been examined in the Siemens Elmiskop I operating at 80 kV.

RESULTS

The microstructure of a 0·14%C steel water-quenched from 950° C and subsequently tempered for 1 h at 100° C is given in Fig. 2. The plate-like precipitates present are 40–80 Å wide and 600–1 200 Å long. In general these precipitates are formed on three clearly defined crystallographic directions, but there are some particles which depart considerably from the common orientations. The crystal structure of the precipitates has yet to be determined. Tempering the same steel for 1 h at 400° C produces a similar microstructure. In this case the precipitates, which are 100–250 Å wide and 600–1 500 Å long, are all arranged in a definite Widmanstätten pattern and have been identified as cementite by electron diffraction.

The martensitic structure produced by water-quenching a 0·98%C steel from 850° C showed light areas outlining the martensite needles. These are due to grain-boundary etching during electropolishing and are useful in identifying the individual needles.

Certain of the martensite needles show striations extending diagonally across the grains. Examination of these striations at high magnification (Fig. 3) shows that some of the wider bands consist of a number of thin bands, the minimum width of a single band being about 20 Å. Greninger and Troiano[10] have proposed a two-stage mechanism for the martensite transformation. The first stage consists of a homogeneous shear on the habit plane producing the well known relief effects, while the second stage is a shear, heterogeneous on a macroscopic scale, but homogeneous within narrow lamellæ about 18 atomic planes in thickness. This second shear causes no overall change in shape of the martensite needle, but should give rise to lamellæ some 20 Å wide running diagonally across a martensite grain. This prediction agrees very well with the results obtained.

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The mechanical properties of hardened high-speed tool steels

A REVIEW OF METHODS OF DETERMINATION

By J. Clayton-Cave and E. Ineson, B.Sc., F.I.M.

SYNOPSIS

An appraisal is given of the methods at present available for testing hardened tool steels in an attempt to evaluate the property of 'toughness'. 1716

HIGH-SPEED STEELS are very complex materials and although much has been done in recent years to elucidate their constitution and the nature of the structural changes which can occur in them, the effects of these changes on their service has by no means been fully explored. This is partly due to the difficulty of obtaining reliable information on the properties of hardened tool steels.

The ultimate guide to the quality of these steels and the effectiveness of their heat treatment is obtained when they are used in the form of tools to do work on some other material. Even so, unless very special precautions are taken to eliminate or control the various factors contributing to the working conditions, it is not easy to arrive at a performance which refers to the tool itself rather than to the conditions under which it is being used.

Although the difficulties of testing tool steels by cutting tests could be overcome, the method of testing is tedious and wasteful, and in any event tends to give an overall appraisal of the tool rather than a measure of the individual properties contributing to its performance. Hence, for the purpose of a research on tool steels it seems preferable in the first place to examine each individual property separately with respect to the effect of variations in manufacture and processing of the steels. This obviously involves recognition of what are the important properties contributing to the overall efficiency of a tool steel,

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followed by the development of adequate tests for detecting even small differences in the value of each property.

PROPERTIES OF HARDENED HIGH-SPEED TOOL STEELS

Normally the only property ever examined by users of high-speed steels is that of hardness. The standard test employed for the determination of this property is that of the diamond indentation test carried out at room temperature. As one might expect, therefore, there is much published literature containing data on room temperature hardness values obtainable with variations in composition, heat treatment, etc. Roberts, Grobe, and Moersch[1] gave details of a very comprehensive study of the hardness of tool steels in various conditions of heat treatment.

It has often been claimed in the literature that the hardness of high-speed steel at room temperature is not so important as that at elevated temperatures. Supporting this point of view, Harden and Grove,[2] and later Bishop and Cohen[3] described methods of determining hardness at elevated temperatures. These techniques have been used to some extent for following the changes in hardness of high-speed steel with variation of temperatures.

Stotz[4] has questioned the real importance of ' hot hardness ' and certainly for many present-day engineering applications where high-speed steels are used to a great extent, the value of hot hardness is difficult to appreciate. This is particularly so for such intermittent cutting operations as milling or hobbing, where, in addition to having only momentary contact between cutting edge and workpiece, there is normally a copious flow of cutting fluid directed at the seat of cutting. Undoubtedly for very heavy rough machining operations the property of ' red-hardness ' as originally demonstrated by Taylor and White (whose work was reported by Becker[5]) is important, and it is not unreasonable to consider that the more precision continuous-cutting operations, such as turning or drilling, would require a tool having resistance to softening at elevated temperatures.

Usually ' rubbing back ' and wear of the cutting edge is associated with softening, and yet these symptoms of failure are not uncommon under conditions of service which would not normally be expected to give rise to excessive heat generation due to frictional forces at the tool point. In intermittent cutting, it must be remembered that in addition to the normal forces, an impact takes place between the cutting edge and the workpiece with each individual stroke. Thus it is not difficult to envisage that initiation of the failure is due to minute chipping of the cutting edge, as suggested by Stotz and others. A more gross type of chipping is in any event a common type of failure of cutting edges.

The importance of investigating the properties of tool steels which may effect the susceptibility of the material to failure of this type has been realized for a number of years and there is considerable reference in the literature to ' toughness ' of tool steels in this connection.

As Jeffries and Archer [13] pointed out, toughness is a complex property. They considered that strength and ductility constitute essential components of toughness. This view has considerable support although the true value implied by the factor ' strength ' and whether this incorporates the hardness value of the material is not always clear. It might be argued that the value of hardness is implied in the strength component, since to some extent the hardness determines the strength, but this dependence has by no means been established in quantitative terms.

Materials like hardened tool steels are generally associated with low toughness. This is understandable if ductility is a component of the toughness, since such materials normally exhibit little or no deformation before failure in the mechanical tests. Any test, therefore, which is to be employed in obtaining a measure of toughness must be extremely sensitive to variations in deformation of specimens under test.

There have been numerous attempts to define ' toughness ' in terms which would permit mathematical treatment to be applied to determine its value, and much work has been undertaken with a view to establishing methods of rendering it directly measurable.

REVIEW OF METHODS EMPLOYED TO INVESTIGATE TOUGHNESS OF TOOL STEELS

Static torsion test

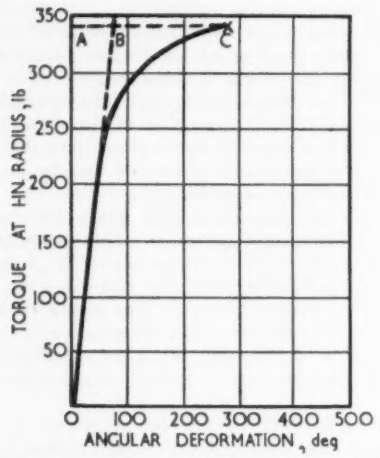
Although torsion tests in various forms have long been used to study the properties of steels, the first important reference to the use of this test for examining hardened tool steels and similar hard materials is that reported by Emmons[6-8] who developed a static torsion test for this purpose.

In the tests employed by Emmons, torque was applied to the specimen in increments of 25 lb/in. and for each increment the angular deformation in degrees was recorded. From the results obtained, torque-deformation curves were plotted, a typical curve for an unspecified steel is shown in Fig. 1. It was found that for some specimens initial failure started in tension, often proceeding with great violence, whilst in others initial failure was in shear. No special significance appears to have been attached to the manner in which the specimens failed, even though it is evident that failure was occurring by two entirely different mechanisms. Further, it is not clear whether all results were used in obtaining the torque-deformation curve or whether those failing in tension were omitted. It is known that in the torsion test the ratio of the maximum normal stress to the maximum shear stress is one to one, and in consequence an equal number of failures through brittle fracture as through ductile fracture might be expected, all external factors being equal. Some of the external factors which might influence the type of fracture obtained are:

(i) misalignment of the specimen
(ii) the presence of residual stresses in the specimen
(iii) the presence of any surface defect causing a stress concentration.

Of these only the latter appears to have been considered by Emmons.

Having obtained a torque/deformation curve, Emmons considered that the total deformation as



A–C total deformation; A–B elastic deformation B–C plastic deformation

Fig. 1—Typical stress/strain curve obtained in Emmons static torsion test, showing method of obtaining plastic deformation

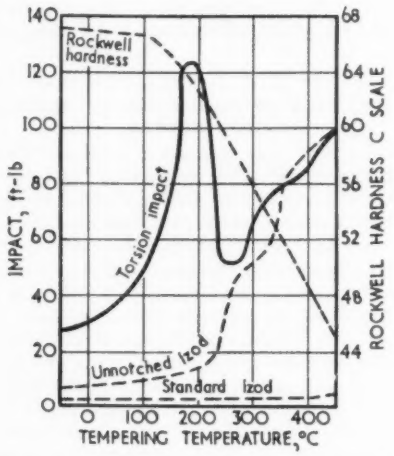


Fig. 2—Relationship between Izod and torsion impact tests on 1·06% plain carbon steel

indicated by the angular deformation in degrees could be divided into the elastic deformation and the plastic deformation at the ultimate torque as indicated in Fig. 1. As a measure of toughness Emmons believed that the product of ultimate torque and total deformation, which he termed ' coefficient of toughness,' gave at least a rough indication of the property. It will be realized that this value is greater than the value of total work to cause failure, sometimes taken as a measure of toughness, by an amount equal to the area OAC in Fig. 1. The reason for the inclusion of the area OAC in the ' coefficient of toughness ' is not easy to understand, since in the case of hardened tool steels where small differences at low levels of toughness are under consideration, these differences could quite easily be masked by a small change in the elastic behaviour of the material when utilizing the Emmons method of assessing toughness.

For the particular test specimen and method used by Emmons it was found that the product of the ultimate torque in pounds at 1 in. radius and the total angular deformation in degrees, combined the strength and deformation in about equal proportions for the case of very strong and tough steels.

Torsion impact test

Luersson, Greene, and Stout[9–11, 16, 18] devoted considerable effort to developing a torsion impact test for the purpose of studying hardened tool steels. Their method differed from that of Emmons in that a momentum unit, consisting of a pair of balanced flywheels, was employed to give a torsional impact to the specimen. The energy absorbed in breaking the test pieces, measured in foot-pounds, was regarded as a measure of toughness. They claimed that the test was capable of good reproducibility particularly if care and attention were given to the preparation of the test pieces.

Exploring the value of the test, direct comparisons were made with conventional impact methods, and for a 1% plain carbon steel it appeared that the torsion impact could indicate changes in resistance to impact which were not detected by the more usual type of impact test (*see* Fig. 2). Considering this behaviour in the light of the dynamics of the test it was concluded that a study of ductility was necessary to explain fully the nature of the torsion impact curve. An empirical formula was evolved to show the relationship between total work to rupture and permanent angle of twist.

Luersson and his co-workers tried to obtain values for ductility by reassembling the broken pieces after test and measuring the permanent deformation of the test pieces. This was possible in some instances where failure occurred in shear, but as in Emmon's static torsion experiments, many test pieces shattered violently and reassembly was impossible. Thus direct measurement of ductility in the torsion impact test is impracticable. Later work by Knight[19] using an improved design of machine confirmed this fact. In one series of experiments involving 36 torsion impact tests, only two values could be obtained for permanent angle of twist.

Green and Stout[18] developed a mathematical method of analysing the stress/strain characteristics of the torsion impact test using a so-called ' sub-rupture ' test to provide values of ductility. In this test the energy of the flywheel was limited to low values sufficient to cause twist of the specimens without leading to rupture. A relationship between degree of twist and work required to produce the twist was obtained from which the torque/strain curves could be constructed. These curves were used to derive stress/strain curves which gave a clear picture of the strength and ductility components of the torsion impact test. Concurrently, Green and Stout also derived stress/strain relationships from static torsion tests. The agreement between these curves and those from the torsion impact test confirmed the same workers earlier opinion[16] that the torsion impact test is purely a rapid static test. If this is so, and Mann[17] has asserted that all impact tests below a critical velocity are really static in

nature, there seems to be little purpose in using a dynamic test, particularly as the static torsion test gives more direct information, e.g. concerning ductility, than is normally possible with the torsion impact test.

Conventional impact tests

Standard notched impact tests of the Izod or Charpy type have been found of little value in comparing hardened tool steels, since in general they are very notch sensitive. Unnotched impact tests have, however, been widely used although few results have been published. Of these only the work of Grobe and Roberts[23] is worthy of attention. These workers realized from the outset that the unnotched impact test is insensitive and erratic and further, like the torsion impact test, gives no information regarding the relative values of strength and ductility, which they and others have considered necessary for a proper evaluation of toughness.

It was found that the test is relatively insensitive to changes in hardness between Rockwell C57 and C67, but that it was rather more sensitive to differences in heat treatment and microstructure. Surprisingly enough, no reference was made to the nature of the fractures, and the 'toss factor'[12] appears to have been ignored. It seems likely that the energy of the broken pieces could be a large part of the energy measured and might mask entirely any small difference in impact resistance.

In view of the rather poor discrimination of the impact test for hard materials it is difficult to understand why it has so often been used as the basis of comparison for other methods of test.

Repeated blow test

As referred to earlier, Greene and Stout[18] employed a repeated torsion-impact test with gradual increments of energy to enable them to derive the stress/strain characteristics of the test. They did not, however, use this technique to investigate directly the properties of tool steels in various conditions.

Repeated blow tests with increasing energy have been carried out in pendulum-type impact machine by a number of laboratories. Carr[21] refers to a knife-edge toughness tester used by the Carpenter Steel Co.[24] but little has been published on the application of this tester. It is interesting to note, however, that at energy values below that required to break the cutting edge the extent of denting of the edge can be observed. The actual measure of toughness is taken as the energy required to break the cutting edge and apparently no account is taken of the behaviour of the cutting edge before that point.

Bancroft and Wight[20] tested 0·15-in. dia. specimens in a similar machine to that of the Carpenter Steel Co. and considered that a realistic evaluation of toughness was obtained by determining the maximum blow, measured in g-cm, that a specimen will withstand before breaking. Various treatments and conditions of high-speed steel were investigated using this measure of toughness and confirmation of results put forward by other workers. Apparently no attempt was made to measure the deformation resulting from each blow and hence there is no information available concerning the elastic and plastic behaviour of specimens tested.

Fatigue test

Comparatively recently reference has been made[34] to fatigue testing of ball-race steels and high-speed steels. In the latter case, three casts were examined and fatigue limits of $\pm$ 44·0, $\pm$ 44·5, and $\pm$ 33 tons/in^2 were found at 10^8 cycles. This result was found to be in good agreement with the microstructures, the poor cast showing larger grain size, worse carbide distribution and more angular carbide particles than the other two casts. The poor cast was of higher vanadium content than the others.

Bend test

Many years ago Barry[15, 25] used a slow bend test to measure the toughness of hardened tool steel and even attempted to compare his test with a Charpy impact test using a small 5-lb pendulum. The work was not, however, very extensive. D'Arcambal,[26] Amberg[27] and others have also used bend tests, but in recent years Grobe and Roberts[22] devoted more detailed thought to the use of such a test for high-speed steel. They developed a bend test using a four-point loading system which, whilst subject to some degree of scatter, seemed capable of giving reasonably accurate measurements of strength and ductility of hardened tool steel. They reported values for yield strength, bend strength, total deflection, and plastic deflection. A true stress/strain diagram for fully hardened high-speed steel was obtained by using a resistance strain gauge to calibrate the deflection in terms of true strain. Yield strength and plastic deflection were considered to be determining factors in toughness. An instance was quoted where, in examining the effect of hardening temperature of high-speed steel over the range 1 095–1 290° C, the bend strength varied only 7·5% (i.e. between 595 000 and 635 000 lb/in^2) whereas there was a 55% increase in yield strength (from 310 000 to 480 000 lb/in^2) and a 93% reduction in plastic deflection (from 0·108 to 0·008 in.). These values were for a specimen size 0·140 in. by 0·50 in. by 2 in. Grobe and Roberts emphasized the importance of stating the specimen size when reporting bend test results. The nature of the fractures obtained was not referred to.

DISCUSSION OF METHODS OF TESTING HARDENED TOOL STEELS

It is clear that opinions differ on what properties should be considered as contributing to the toughness of a hardened tool steel. As pointed out by Rose,[31] the term 'toughness' is somewhat general having no satisfactory definition and it might perhaps be wise to refer to more well established terms which have a definite meaning.

It still remains to decide what properties are desirable in a hardened tool steel to enable it to withstand the stresses imposed upon it when in use as a cutting tool. Presumably it is some property or combination of properties which gives good resistance to breaking, whether this is in the form of minute chipping at the cutting edge or of a more gross character. The majority

of previous workers have considered that one important factor is ductility, i.e. the ability to deform before breaking. The torsion test, particularly the static torsion test, and the bend test have been shown capable of measuring ductility and in work with these machines much emphasis has been given to values of plastic deformation. This is difficult to understand, since, as pointed out by Zuilen,[28] plastic flow or deformation can rarely be tolerated in precision cutting tools. It would seem that the elastic properties are of greater significance, i.e. that a comparison of the energies which can be absorbed elastically by materials in various conditions would give a truer evaluation of so-called 'toughness'.

Grobe and Roberts[22] did evaluate yield strengths from their bend tests, although on studying their stress/deflection or stress/strain curves it is not easy to see how they could be so precise in selecting values. They did not, however, consider the elastic deformation separately from the total deformation and they considered that as there appeared to be no measurable difference in Young's modulus the yield point must be the only significant value.

More recently, Hamaker, Strang and Roberts[33] have studied bend/tensile relationships for high-tensile steels, and have produced correlation curves for bend and tensile yield strengths and fracture stresses. The same workers also observed a relationship between hardness and Young's modulus measured in tension.

Whether there is any variation in Young's modulus appears to be controversial and because of the lack of definite information regarding elastic properties of hardened tool steels a careful study might well prove interesting. In carrying out tests a suitable bend test has much to commend it since the specimen is of simple geometric form, comparatively easy to machine and heat treat, and the test can be performed without undue difficulty. The torsion test would be likely to give similar results, but it offers no advantage over the more simple bend test and suffers from the disadvantage of using a test piece which is difficult to prepare. In any event, the great majority of types of tools are not subject to torsional stresses and it is suggested that torsion tests should be reserved for applications where they are of particular interest, e.g. for drills.

A repeated blow test might well be worthy of consideration especially if some device could be introduced to enable readings of deformation to be made. It should at least be possible to determine the blow at which permanent deformation first appears and in any event it would give a quick and easy method of determining the energy necessary to rupture the material. A test of this type might be particularly suitable for routine works testing if it could be shown to give a reliable evaluation of tool steels.

Other mechanical tests

Having advanced the idea that the ability to absorb energy without yielding is one of the important properties of hardened tool steels, other tests not previously used to any extent in this field suggest themselves.

Practically, under stresses approaching service conditions, fatigue tests might be carried out to determine how tool steels, heat-treated in various ways, will withstand alternating stresses at various stress levels. Such tests are likely to present considerable practical difficulties and could not be undertaken lightly, but it is felt that the results might prove most interesting and of great practical value.

More fundamentally, measurements of damping capacity seem worthy of attention, and making such measurements does not present great practical difficulties to the research worker. The technique might prove to be of particular use in studying the effects of variations in microstructure, grain size, carbide size and distribution, internal stresses, etc.

Other tests to be correlated with main mechanical tests

In attempting to obtain good resistance to breaking, the result must be consistent with adequate hardness and wear resistance for any given duty. Hence there is little purpose in investigating variations of breaking resistance at hardness below those normally acceptable, e.g. below, say, 800–850 VPN. Frequently, differences in performance are observable between tools of the same apparent hardness and yet little attention has been given to determining differences in mechanical properties of tool steels heat-treated by various means to an equal final hardness. Grobe and Roberts[22] refer briefly to this, but insufficient data were given to reach any definite opinion on the matter.

In reporting values of hardness, almost all previous workers have given the macro-hardness which is presumably some average of carbide and matrix. A few investigators, including Tarasov,[29] have reported hardness values for the carbide constituents, but little attention has been given to determining the hardness of the matrix. Bergsman[30] has given an isolated example of hardness tests on the matrix of a hardened high-speed steel which is sufficient to illustrate the possibilities. There are no doubt practical difficulties in making such measurements but it is felt that the matrix hardness is of much more direct concern when making comparisons with other properties.

Another technique which has come into prominence recently and warrants consideration is the measurement of X-ray line broadening for studying the stability of hardened tool steels. This has been described by Wainwright and Nickols.[32] The effect of tempering in the range 100–150° C might be investigated with this technique and correlated with any differences observed in other properties.

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The spectrographic determination of residual elements in steel

REPORT No. 2 THE EXCITATION INDEX TECHNIQUE

By the BISRA Spectrographic Analysis Sub-Committee

SYNOPSIS

A method has been developed for the determination of the commoner residual elements in mild- and low-alloy steels. It is applicable to a variety of intermittent ac and dc arc source units. The excitation conditions are specified by means of the intensity ratio of a non-homologous line pair of the major element. The method has been applied to the determination of cobalt, nickel, titanium, copper, zirconium, tin, molybdenum, niobium, vanadium, aluminium, and chromium at contents down to 0·01%, and a mean reproducibility, in terms of coefficient of variation, of ± 4·2% has been achieved. 1703

THE FIRST REPORT[1] of the study group of the Spectrographic Analysis Sub-Committee* described the determination of a limited number of residual elements using intermittent ac arc excitation. In this report it was suggested that although a standard design of source unit might readily enable a standard method to be formulated, individual laboratories would not be prepared to abandon satisfactory non-standard equipment solely for the purpose of greater uniformity of method. This view was subsequently confirmed as changes in membership of the Group brought about a corresponding change in the excitation equipment available. All members had source

* The constitution of the Sub-Committee at 31st December, 1958, was: Mr J. H. Oldfield, Chairman, (Admiralty Materials Laboratory), Mr A. Argyle (British Cast Iron Research Association), Mr F. R. Bareham (British Non-Ferrous Metals Research Association), Mr K. M. Bills (Mond Nickel Co. Ltd), Mr P. Gale (Steel Company of Wales Ltd), Mr M. Z. de Lippa (Consett Iron Co. Ltd), Mr G. S. Marsden, (Edgar Allen and Co. Ltd), Mr D. F. Sermin (Samuel Osborn and Co. Ltd), Mr H. T. Shirley (Brown-Firth Research Laboratories), Mr W. S. Sykes (Steel, Peech and Tozer), Mr F. van Someren (Murex Welding Processes Ltd). The members of the Study Group were: Mr K. M. Bills, Mr H. J. Dothie, replaced by Mr C. Park, (Government Chemists Laboratory), Mr D. F. Sermin, Mr S. R. Shutt (Bragg Laboratory, Noie), and Mr R. H. Tyas (BISRA).

Paper MG/DD/272/58 of the Spectrographic Analysis Sub-Committee of the Metallurgy (General) Division of the British Iron and Steel Research Association, received on 4th March, 1959.

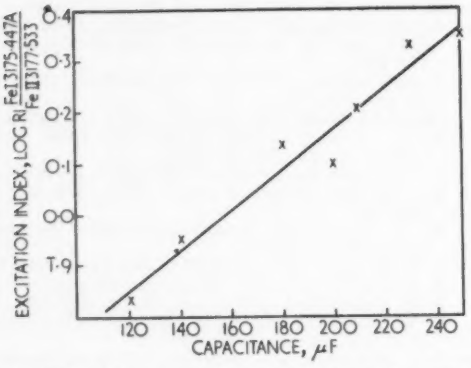


Fig. 1—Effect of capacitance on excitation index, Hilger–BNF unit

units of the triggered low-voltage arc type, that is, variants of the original Pfeilsticker circuit, but the units differed considerably in detail; indeed no two were alike. Both intermittent ac and dc arcs were represented, as follows:

Laboratory	Source type	Special characteristics
A	dc	Hilger-BNF general purpose unit
B	ac	Sinclair[4] type, trigger auto-transformer-coupled
C	ac	Kingsbury and McLelland[3] type, auto-transformer-coupled trigger
D	dc	Multisource type electronic trigger
E	ac	Kingsbury and McLelland[3] type, Tesla coupled trigger.

The outstanding practical difference between the two dc source units was that the trigger of the Hilger–BNF unit contributed a significant part of the total light output, whereas the contribution of laboratory D's electronic trigger was virtually negligible. It was felt that this excitation equipment represented a reasonable cross-section of such equipment at present employed in the UK.

The terms of reference of the study group were to devise a method for the determination of the commoner residual elements in mild- and low-alloy steels, having a reproducibility of better than $\pm 10\%$ and limits of detection of 0·01% or better. The method was to be limited to the analysis of solid samples.

For the method to be compatible with a wide variety of source units it was necessary to ensure that all units produced similar spectra. The major task of the group, therefore, resolved itself into a study of the various factors affecting the spectrum characteristics, and a search for some practical numerical means of defining them. Two possible approaches were examined:

(i) Precise specification of the discharge current waveform, peak current, r.m.s. current and time constant. Such a method is essentially indirect, inasmuch as it measures cause rather than effect.

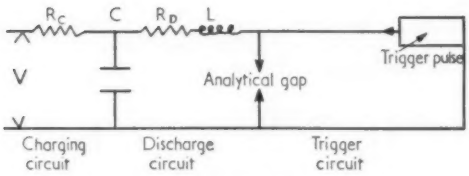
TABLE I

Excitation index test

Analysis line pair Å		Calibration graph index value, %				
		Laboratory				
		A	B	C	D	E
Co I	3453·505					
Fe I	3459·918	0·073	0·080	0·040	0·105	0·109
Ni I	3414·765					
Fe I	3415·534	0·047	0·048	0·025	0·059	0·049
Ti II	3349·406					
Fe	3355·229	0·020	0·024	0·026	0·025	0·018
Cu I	3273·962					
Fe	3280·261	0·030	...	<0·01	0·033	0·023
Zr II	3273·047					
Fe II	3281·300	0·038	0·023	0·042	0·025	0·022
Sn I	3262·328					
Fe I	3253·602	0·070	0·063	0·038	0·068	0·084
Mo	3170·347					
Fe II	3171·353	0·020	0·028	0·017	0·016	0·017
Nb	3130·786					
Fe II	3135·364	0·110	0·080	0·020	0·074	0·063
V II	3093·108					
Fe	3093·357	0·011	0·036	0·015	0·012	0·009
Al I	3082·155					
Fe	3080·112	0·027	0·027	0·039	0·024	0·018
Cr II	2849·838					
Fe II	2849·606	0·19	0·107	0·179	0·155	0·096
		Excitation index, Log RI				
Fe I	3175·447					
Fe II	3177·535	0·24	0·29	0·25	0·26	0·25

(ii) Direct assessment of the degree of excitation from the actual spectra; a numerical expression of the appearance can be obtained by measurement of the ratio of the intensities of two lines of the major element with widely differing excitation potentials. This ratio has commonly been described as 'excitation index'.

It was known from earlier work that both of these methods were capable of achieving a high degree of uniformity when applied to source units of similar electrical design. There was, however, considerable doubt whether the first (electrical) method could be used to match units operating at different frequencies. The intermittent ac arc units delivered 100 discharges/second, which was twice the frequency of the dc units. Furthermore, an experiment by laboratory A showed that virtually identical spectra could be produced by the same unit using discharges of widely different discharge current waveforms. The relevant circuit parameters investigated in this particular test were, (*a*) 120 μF 11Ω 60 μH, and (*b*) 170 μF 50Ω 60 μH. The excitation index method was more attractive in that its numerical expression was immediately intelligible to the spectrographer and that no special equipment was required for its measurement. There were, however, no grounds for assuming that the intensities of lines of the residual elements themselves were simply related to the excitation index, but such a hypothesis was a good basis for a collaborative investigation.



Schematic circuit diagram

Fig. 2—Excitation and exposure conditions used for the excitation index and reproducibility tests

Laboratory	V	R_C, Ω	C, microfarad	R_D, Ω	L, millihenries	Upper electrode*	Analytical gap spacing, mm	Exposure time, s
A	350 d.c.	...	230	15·0	0·06	90° pointed cone	2·0	10
B	230 a.c.	40	80	6·7	...	90° pointed cone	3·0	25
C	230 a.c.	30	40	7·0	0·5	80° cone 1-mm radius ball tip	2·0	50
D	350 d.c.	...	180	5·1	1·0	90° cone 1-mm radius ball tip	2·5	40
E	200 a.c.	60	48	...	...	,,	2·0	120

*All laboratories used ¼-in. dia. graphite lower electrode, sample linished flat, 60 grit

EXPERIMENTAL PROCEDURE

The following details of technique were common to all laboratories:

Spectrograph	Hilger, large quartz Littrow
Slit width	15μ
Wavelength range	3 500–2 450 Å
Upper electrode	¼-in. dia. graphite rod
Photographic plate	Kodak B 10.

Plate calibration in terms of relative intensities was used by all members.

In all other details members followed the practices normal in their own laboratories. This often involved considerable variations, particularly in counter electrode shape, analytical gap spacing, and photographic processing. Since the key to the whole investigation was the validity or otherwise of the excitation index/spectrum characteristic hypothesis, a deliberate effort was made to avoid an excessive amount of standardization. The Bureau of Analysed Samples Spectrographic Standards (Mild Steel Residual Series) SS 11–17 were used for calibration purposes.

INVESTIGATION OF THE EXCITATION INDEX

A number of pairs of iron lines with widely different excitation potentials were examined to determine their suitability as a means of measuring the degree of excitation in the spectra. The line pair finally chosen was

$$\frac{\text{Fe I } 3175{\cdot}447 \text{ Å}}{\text{Fe II } 3177{\cdot}535 \text{ Å}}$$

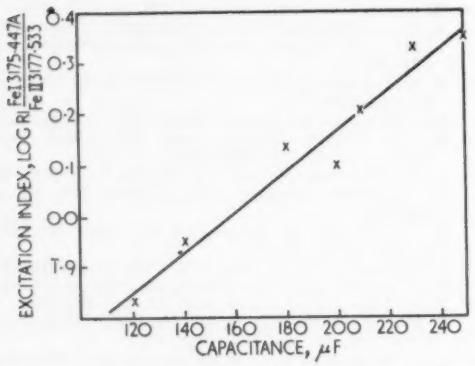
and the logarithm of the relative intensity ratio was defined as the excitation index. An increase in this value indicates a more arc-like discharge and vice versa.

A preliminary survey of the significance of this index was carried out in which standards were exposed by the technique normally employed by each laboratory. Measurements were made of the excitation indices and it was observed that these differed markedly between the various standards. SS 11 was therefore arbitrarily chosen as the reference standard. Visual comparison of the spectra indicated that those produced by the ac units were much more dense and arc-like than those of the dc units. These observations were reflected in the actual measurements; the ac units produced excitation index values of (log) 0·4–0·6 while both dc units gave values of about $\bar{1}{\cdot}8$.

The apparent relationship between excitation index and sensitivity was, however, encouraging and a second test was carried out in which all members tried to achieve an excitation index of 0·0 at a density level of 0·9 (12–14% transmission). The electrical parameters of the discharge circuits were altered to bring this about; time constants were increased and/or peak current decreased to make the discharge more arc-like and vice versa. A typical response to such adjustments is shown in Fig. 1 which was recorded by laboratory A (Hilger-BNF unit) on varying the discharge capacitance.

The two intermittent dc units successfully achieved the required excitation index value, but it was found to be impossible to reduce the sensitivity of the ac units sufficiently. A marked improvement in uniformity between spectra from different units was, however, obtained and this improvement had obviously been brought about by the control of excitation index and density level. A compromise value of 0·3 for the excitation index was finally found to be within the range of all source units.

Having thus reached a critical stage of the investigation the choice of analysis line pairs was considered and a suitable list finally approved. Tests were also conducted on the reproducibility of the excitation index. This was found to be very low, as might be expected from a line pair deliberately selected to be



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V II	3093·108					
Fe	3093·357	0·011	0·036	0·015	0·012	0·009
Al I	3082·155					
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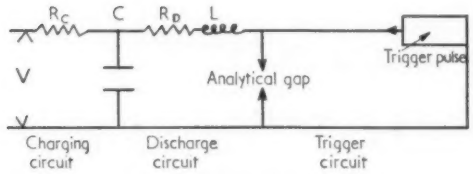
Fe II	3177·535	0·24	0·29	0·25	0·26	0·25

(ii) Direct assessment of the degree of excitation from the actual spectra; a numerical expression of the appearance can be obtained by measurement of the ratio of the intensities of two lines of the major element with widely differing excitation potentials. This ratio has commonly been described as 'excitation index'.

It was known from earlier work that both of these methods were capable of achieving a high degree of uniformity when applied to source units of similar electrical design. There was, however, considerable doubt whether the first (electrical) method could be used to match units operating at different frequencies. The intermittent ac arc units delivered 100 discharges/second, which was twice the frequency of the dc units. Furthermore, an experiment by laboratory A showed that virtually identical spectra could be produced by the same unit using discharges of widely different discharge current waveforms. The relevant circuit parameters investigated in this particular test were, (*a*) 120 μF 11Ω 60 μH, and (*b*) 170 μF 50Ω 60 μH. The excitation index method was more attractive in that its numerical expression was immediately intelligible to the spectrographer and that no special equipment was required for its measurement. There were, however, no grounds for assuming that the intensities of lines of the residual elements themselves were simply related to the excitation index, but such a hypothesis was a good basis for a collaborative investigation.



Schematic circuit diagram

Fig. 2—Excitation and exposure conditions used for the excitation index and reproducibility tests

Laboratory	V	R_C, Ω	C, micro-farad	R_D, Ω	L, milli-henries	Upper electrode*	Analytical cap spacing, mm	Exposure time, s
A	350 d.c.	...	230	15·0	0·06	90° pointed cone	2·0	10
B	230 a.c.	40	80	6·7	...	90° pointed cone	3·0	25
C	230 a.c.	30	40	7·0	0·5	80° cone 1-mm radius ball tip	2·0	50
D	350 d.c.	...	180	5·1	1·0	90° cone 1-mm radius ball tip	2·5	40
E	200 a.c.	60	48	...	...	,,	2·0	120

*All laboratories used ¼-in. dia. graphite lower electrode, sample linished flat, 60 grit

EXPERIMENTAL PROCEDURE

The following details of technique were common to all laboratories:

Spectrograph	Hilger, large quartz Littrow
Slit width	15μ
Wavelength range	3 500–2 450 Å
Upper electrode	¼-in. dia. graphite rod
Photographic plate	Kodak B 10.

Plate calibration in terms of relative intensities was used by all members.

In all other details members followed the practices normal in their own laboratories. This often involved considerable variations, particularly in counter electrode shape, analytical gap spacing, and photographic processing. Since the key to the whole investigation was the validity or otherwise of the excitation index/spectrum characteristic hypothesis, a deliberate effort was made to avoid an excessive amount of standardization. The Bureau of Analysed Samples Spectrographic Standards (Mild Steel Residual Series) SS 11–17 were used for calibration purposes.

INVESTIGATION OF THE EXCITATION INDEX

A number of pairs of iron lines with widely different excitation potentials were examined to determine their suitability as a means of measuring the degree of excitation in the spectra. The line pair finally chosen was

$$\frac{\text{Fe I } 3175{\cdot}447 \text{ Å}}{\text{Fe II } 3177{\cdot}535 \text{ Å}}$$

and the logarithm of the relative intensity ratio was defined as the excitation index. An increase in this value indicates a more arc-like discharge and vice versa.

A preliminary survey of the significance of this index was carried out in which standards were exposed by the technique normally employed by each laboratory. Measurements were made of the excitation indices and it was observed that these differed markedly between the various standards. SS 11 was therefore arbitrarily chosen as the reference standard. Visual comparison of the spectra indicated that those produced by the ac units were much more dense and arc-like than those of the dc units. These observations were reflected in the actual measurements; the ac units produced excitation index values of (log) 0·4–0·6 while both dc units gave values of about $\bar{1}$·8.

The apparent relationship between excitation index and sensitivity was, however, encouraging and a second test was carried out in which all members tried to achieve an excitation index of 0·0 at a density level of 0·9 (12–14% transmission). The electrical parameters of the discharge circuits were altered to bring this about; time constants were increased and/or peak current decreased to make the discharge more arc-like and vice versa. A typical response to such adjustments is shown in Fig. 1 which was recorded by laboratory A (Hilger-BNF unit) on varying the discharge capacitance.

The two intermittent dc units successfully achieved the required excitation index value, but it was found to be impossible to reduce the sensitivity of the ac units sufficiently. A marked improvement in uniformity between spectra from different units was, however, obtained and this improvement had obviously been brought about by the control of excitation index and density level. A compromise value of 0·3 for the excitation index was finally found to be within the range of all source units.

Having thus reached a critical stage of the investigation the choice of analysis line pairs was considered and a suitable list finally approved. Tests were also conducted on the reproducibility of the excitation index. This was found to be very low, as might be expected from a line pair deliberately selected to be

TABLE II

Reproducibility test, sample ref. no. MGS/308

Element	Mean content, %					Coefficient of variation, %				
	Laboratory					Laboratory				
	A	B	C	D	E	A	B	C	D	E
Co	0·050	0·050	0·059	0·056	0·050	3·0	3·0	3·4	3·0	2·8
Ni	0·049	0·052	0·050	0·058	0·040	3·1	5·1	3·0	2·8	3·2
Ti	0·050	0·057	0·048	0·057	0·051	3·0	4·0	6·2	2·4	3·7
Cu	0·053	...	0·040	0·049	0·050	3·6	...	5·0	2·8	5·4
Zr	0·024	0·016	0·020	0·018	0·026	5·8	7·8	7·1	5·4	9·0
Sn	0·058	0·069	0·055	0·052	0·059	4·5	4·4	5·4	5·8	5·9
Mo	0·057	0·066	0·057	0·044	0·049	3·2	3·0	5·2	6·1	5·0
Nb	0·047	0·047	0·051	0·047	0·046	4·5	4·7	6·5	4·2	6·0
V	0·058	0·056	0·050	0·049	0·050	3·3	4·5	3·0	2·0	3·1
Al	0·052	0·057	0·050	0·052	0·053	2·1	3·5	3·0	4·0	4·7
Cr	0·051	0·053	0·052	0·056	0·054	2·7	4·2	3·8	2·3	3·8

non-homologous; a standard deviation of $\pm \log 0{\cdot}050$ was recorded. It was also observed that the index increased with increasing exposure time and there was considerable evidence to show that the effect was thermal in origin. A reasonable latitude in comparing excitation indices and corresponding analysis lines was obviously necessary.

Conditions were then defined for a final test of the excitation index/spectrum sensitivity relationship. Standards SS 11–17 were exposed with the various source units adjusted to achieve an excitation index value of 0·2–0·3 at a density level of 0·9 for FeI 3 175·447, as measured on the spectrum of SS 11. The calibration curves were established and limits of detection and index values (i.e. equality points) of each line pair were measured. All laboratories achieved a limit of detection of 0·01% on all elements with the exception of tungsten, where the limit was of the order of 0·05%. The line pairs used and results obtained are recorded in Table I. It will be seen that, within reasonable limits, similar index values were obtained for each analysis line pair. All members encountered difficulty in obtaining a good calibration curve for zirconium.

It was considered that the test fully justified the use of the excitation index at a given density as a means of defining the type of spectrum. It has led to a general uniformity with widely different source units. It remained to be proved whether the excitation index thus defined was capable of obtaining analyses of the required reproducibility.

REPRODUCIBILITY OF THE TECHNIQUE

A sample of mild steel was prepared containing about 0·05% of each of the elements of interest. The material was cast, extensively worked and heat-treated to secure the maximum degree of homogeneity. Each member was supplied with three 1-in. lengths of 1¼-in. dia. bar with the flat surfaces linished to a standard 60 grit finish. Each piece of sample was sparked 10 times with excitation conditions as in the preceding test. Electrical and other parameters used for these tests are given in Fig. 2. Care was taken to spark each piece consecutively to avoid excessive heating effects. The standard deviations of the log relative intensity ratios were calculated and the coefficients of variation were obtained from this value using the calibration curves already established. The results obtained are recorded in Table II. All members succeeded in obtaining coefficients of variation of better than 10% and in general, good agreement in the reported mean contents was achieved.

The reproducibility of the zirconium values was consistently inferior to that of other elements and it was noted that all members had previously encountered difficulty in obtaining good calibration curves for this element. Laboratory B did not report values for copper since the available plate calibration curve did not extend to sufficiently high densities. In general, however, the results were satisfactory.

DISCUSSION

The investigation proved that a good measure of standardization could be achieved with widely different source units by specifying the excitation in terms of the excitation index. A more exact correspondence of the spectra can hardly be expected with such a complex process as the excitation of metal electrodes in air. The excitation index has been shown to increase with time of sparking due, presumably, to thermal effects, but these same effects bring about a decrease in the emission of most residual elements. Unless the spectrographs have exactly the same photographic speed, even identical light sources would not produce identical spectra. The technique evolved in this investigation is not, therefore, proposed as a definitive method for the determination of residual elements in steel, but as providing the basis from which the detailed procedure of such a method can be developed.

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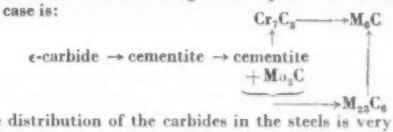
The tempering of 2¼%Cr–1%Mo steel after quenching and normalizing

By R. G. Baker, M.A., Ph.D., and J. Nutting, M.A., B.Sc., Ph.D.

SYNOPSIS

A 2¼%Cr–1%Mo steel has been tempered systematically for times up to 1 000 h in the temperature range 400–750° C after both quenching and cooling at 300° C/h from the austenite range. The hardness changes which occurred have been measured and the accompanying microstructural changes studied. The carbides precipitated were extracted by the direct carbon extraction technique and examined by electron microscopy, X-ray and electron diffraction and by X-ray fluorescence analysis.

The results indicate that the general sequence of carbides formed in each case is:



but the distribution of the carbides in the steels is very different. Mo_2C forms by separate nucleation, but cementite, Cr_7C_3, $M_{23}C_6$ and M_6C form by nucleation at carbide particles already present. In the normalized steel, it was found that precipitation of Mo_2C was not accompanied by any appreciable change in the micro-hardness of the ferrite. The results are discussed in relation to the creep properties of the steel in the quenched and tempered, and normalized and tempered conditions. 1708

A STEEL containing 2¼%Cr–1%Mo is widely used for the production of superheater tubes, and therefore its creep properties have been widely investigated although its metallography has not been studied so intensively. After fabricating into tubes, the steel is normalized and then given a stress-relieving treatment. This latter term is really a misnomer, for pronounced structural changes occur which confer good creep properties upon the steel. In view of this the heat treatment should be referred to as 'tempering' and therefore this term has been adopted throughout the account of this research.

The history of metallurgical investigation shows the value of using a variety of techniques to study a given problem. In addition to the extra information which can often be obtained in this way, there is less risk of the results being misinterpreted. Thus in the present investigation, the precipitated carbides have been studied using four separate techniques. First the carbides were extracted by using the direct carbon replica technique devised by Smith and Nutting[1] and then examined by electron microscopy and electron diffraction, whilst new techniques were developed to enable X-ray diffraction and X-ray fluorescence analysis to be used also. Previous X-ray diffraction and chemical analysis techniques have relied upon the extraction of the carbides by lengthy electrolytic processes. These often give a low recovery, especially in the case of fine carbides. The carbon extraction replica technique has been found to extract all the carbides intersecting the carbon layer and has enabled, therefore, a great increase to be achieved in both the speed and accuracy of the primary extraction process.

Steels of the 2¼%Cr–1%Mo type have a better creep resistance when tempered in the normalized than when tempered in the quenched condition. This paper describes a systematic metallographic investigation of the tempering of a steel of this type after both normalizing and quenching, in an attempt to correlate the different creep properties with the differences in the microstructure. The tempering of steels of this type leads first to the formation of iron carbides and subsequently of alloy carbides. The sequence of carbides formed is somewhat similar on tempering both the quenched and the normalized steels, but the distribution of the carbide particles is very different. Normalizing produces both ferrite and bainite whilst quenching produces auto-tempered martensite. The behaviour of the bainite and martensite is very similar but that of the ferrite is quite different both with respect to hardness and to the sequence of carbides formed. The investigation has been carried out over a range of temperatures between 400° and 750°C for times up to 1 000 h. Particular attention has been paid to the changes in morphology, structure, and chemical composition of the carbides.

The analysis of the steel used and a detailed account of the heat treatments and preparation techniques are given in appendix I.

EFFECT OF TEMPERING ON MICROSTRUCTURES

After quenching

The microstructure of the as-quenched specimen showed acicular martensite grains, many of which contained plate-like precipitates in a Widmanstätten array. Some grains also contained larger plates of lenticular cross-section (*see* Fig. 1). The diffraction pattern from the Widmanstätten precipitates was very weak and diffuse. However from the precipitate morphology and the contributory evidence from X-ray fluorescence analyses it is believed that these precipitates are of ε-carbide. The diffraction patterns from some of the larger lenticular particles gave spots attributable to cementite.

The sequence of carbides formed after the different tempering treatments is shown diagrammatically in Fig. 2. The results are in general agreement with those of other workers on similar materials except

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The authors are with the department of metallurgy, University of Cambridge.

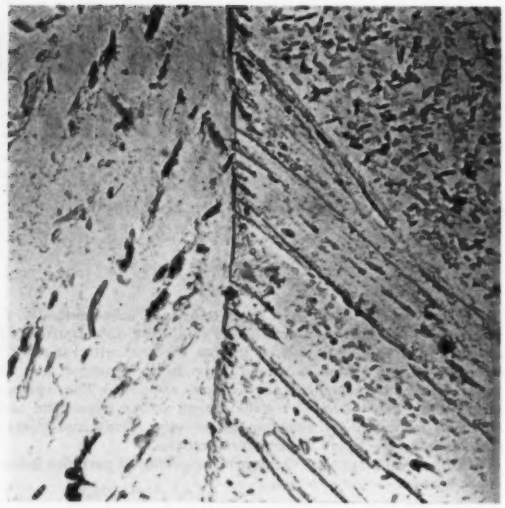


Fig. 1—Cr–Mo steel quenched from 960° C, showing plate-like grains many of which contain Widmanstätten precipitates; carbon extraction replica × 12 500

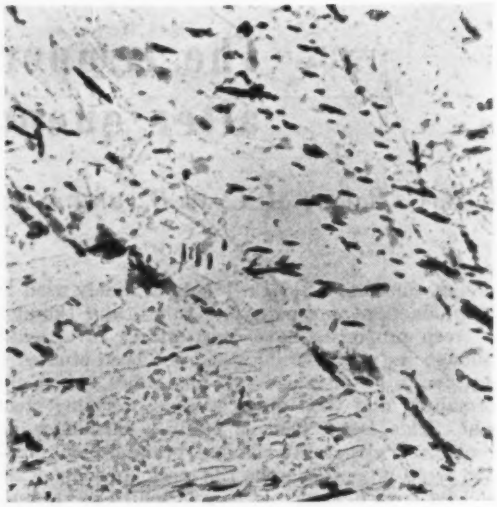


Fig. 3—Cr–Mo steel quenched and tempered for 50 h at 400° C showing growth of the larger lenticular iron carbides; carbon extraction replica × 12 500

that no M_aC_b was detected in this investigation. It is possible that this transitional carbide may have been formed in small quantities but more accurate X-ray techniques would be required to identify it in the presence of $M_{23}C_6$ and M_6C.

Tempering for short times at 400° C resulted in the precipitation of more of the larger lenticular carbides and plate-like carbides at the grain boundaries. There was an apparent orientation relationship between the lenticular carbides and the ferrite matrix since the precipitates in any one ferrite grain were parallel. On further tempering, the carbide precipitates thickened, but the rate of thickening was greatest for those particles situated in the grain boundaries. The lenticular carbides thickened more slowly and the small Widmanstätten precipitates more slowly still. Figure 3 shows the microstructure after tempering for 50 h. The thickening of the plates was accompanied by an increase of the number of spots which could be attributed to cementite in the electron diffraction patterns.

After short times of tempering at 500° C, the microstructure was closely similar to that at 400° C, except that in a few small isolated regions, a heavy precipitate of large branched carbides could be observed (Fig. 4). These have been identified as cementite by using selected area electron diffraction. It is thought

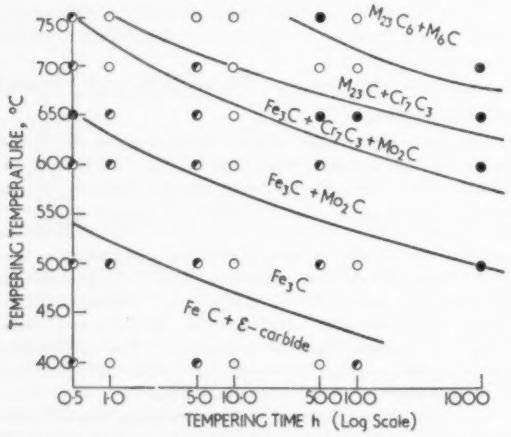


Fig. 2—Isothermal diagram showing the sequence of carbide formation on tempering the quenched steel

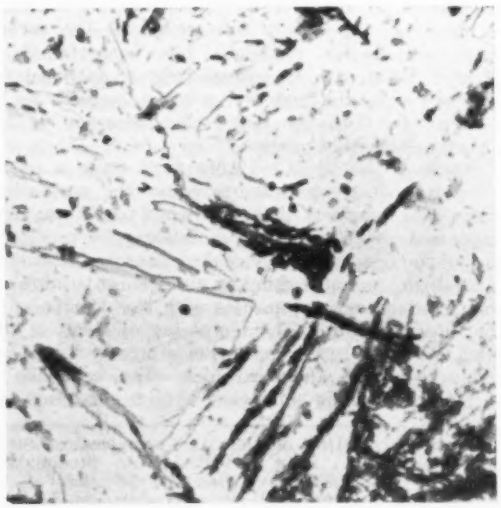


Fig. 4—Cr–Mo steel quenched and tempered for ½ h at 500° C showing dense cementite precipitates at a region of former retained austenite; carbon extraction replica × 12 500

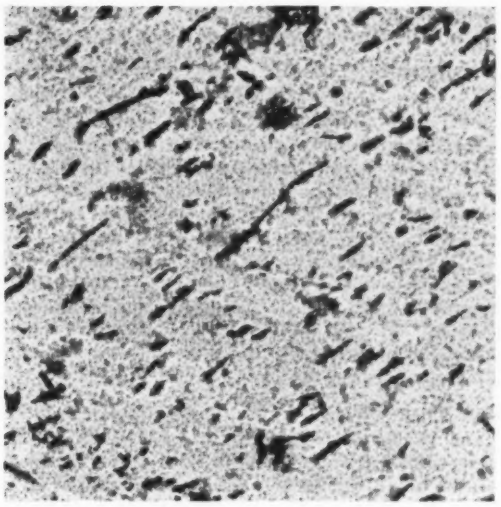


Fig. 5—Cr–Mo steel quenched and tempered for 1 000 h at 500° C showing lenticular cementite and incipient Mo_2C precipitation; carbon extraction replica × 12 500

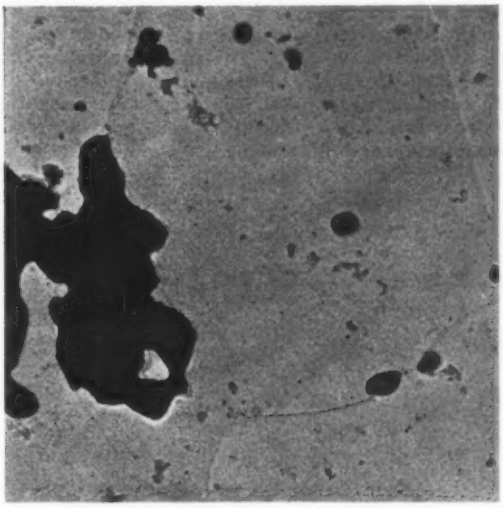


Fig. 7—Cr–Mo steel quenched and tempered for 1 000 h at 700° C showing a cluster of large M_6C particles and smaller $M_{23}C_6$ particles in a matrix of equiaxed ferrite grains; carbon extraction replica × 12 500

that retained austenite was formerly present in these regions and that this decomposed to ferrite and cementite during tempering at 500° C. A relatively high concentration of carbon would have been present in the retained austenite, thus accounting for the large amount of carbide present in these regions.

The precipitation of Mo_2C was preceded by the appearance of small diffuse zones in the ferrite matrix as shown in Fig. 5. On further tempering, small needles were precipitated from the diffuse zone structure in the ferrite and these were indentified as Mo_2C by electron diffraction (*see* Fig. 6). Then cementite spheroidized progressively but there was only slight growth of the Mo_2C needles.

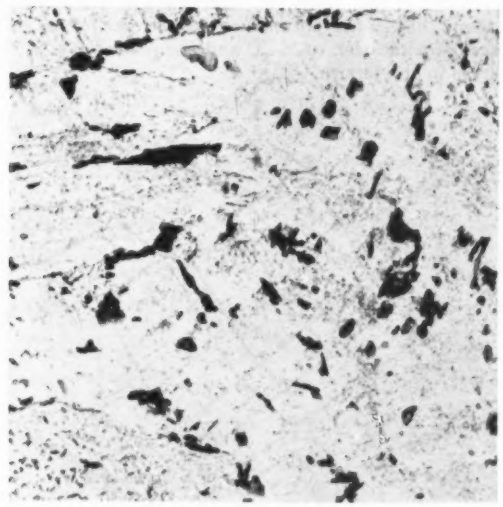


Fig. 6—Cr–Mo steel quenched and tempered for 5 h at 600° C showing the development of small needles of Mo_2C in the matrix; carbon extraction replica × 12 500

The formation of Cr_7C_3 was detected by its X-ray diffraction pattern and its characteristic morphology. It appeared to be precipitated at or within the existing cementite particles; no evidence of its formation by separate nucleation was observed in this steel.

The precipitation of $M_{23}C_6$ took place at 650° C and higher temperatures. It appeared as large, spheroidal particles which were initially found at the prior austenite boundaries. The $M_{23}C_6$ grew at the expense of both the cementite and the Mo_2C leading to a substantial decrease in the number of particles present. This allowed recrystallization of the ferrite matrix, and the newly formed equiaxed grains had $M_{23}C_6$ particles at the grain boundaries.

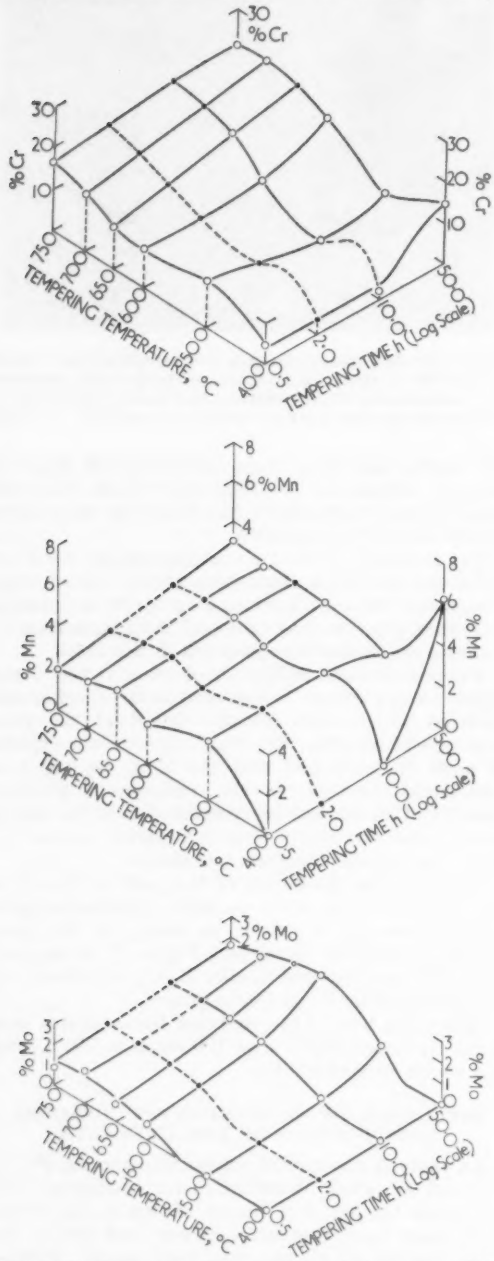
Finally on the formation of M_6C and its growth at the expense of the other carbides, substantial grain growth was able to occur as many of the grain boundary carbides dissolved. Figure 7 shows very large M_6C particles and smaller $M_{23}C_6$ particles at the boundaries of equiaxed ferrite grains.

When first formed the equiaxed ferrite grains were about 1μ diameter but after 100 hours at 750° C they had grown to about 3–4 μ.

INFLUENCE OF TEMPERING ON CHEMICAL COMPOSITION OF THE EXTRACTS

An analysis was carried out for the elements Fe, Cr, Mn, and Mo using W radiation for excitation. The results for Cr, Mn, and Mo are shown in Figs. 8–10. With short tempering times at 400° and 500° C, the total number of counts was very small. Further tempering caused the Cr and Mn contents of the carbides to increase, reach a maximum, and then decrease. These changes were accompanied by an

increase in the total amount of carbide present and are thought to be a consequence of the transformation of ε-carbide to cementite. This transformation will be considered later in the discusssion. At higher tempering temperatures, the Cr content rose considerably whilst a small rise in Mo content corresponding to the precipitation of Mo_2C was found. There was no subsequent fall in Mo content when the Mo_2C dissolved, which indicated the presence of Mo in the $M_{23}C_6$.



Figs. 8–10—Isometric diagrams showing apparent %Cr, Mn, and Mo in the extracted carbides plotted against time and temperature of tempering for the quenched steel

INFLUENCE OF TEMPERING ON VICKERS HARDNESS

The effect of tempering time and temperature upon Vickers hardness is shown in the isometric diagram Fig. 11. In the as-quenched condition the hardness was 419. The hardness decreased slightly on tempering at 400° C but then showed a slow increase with time during the period whilst ε-carbide was transforming to cementite. At 500° C, the hardness decreased slowly with time. There was a maximum in the curve at 600° C corresponding to the precipitation of Mo_2C. At 650°, 700°, and 750° C, there was an almost linear decrease of hardness with the logarithm of the tempering time.

After normalizing

The as-normalized specimen contained both proeutectoid ferrite and bainite; the ratio of ferrite/bainite being about 55 : 45. A micrograph of the structure is shown in Fig. 12. No carbides were present in the ferrite, but plate-like and lath-like carbides were extracted from the bainite regions. These precipitates formed a Widmanstätten pattern and were thought to be partly ε-carbide and partly cementite. In addition, fringes of acicular Mo_2C were observed on the ferrite side of some of the bainite/ferrite boundaries. The individual particles were lath-like ($\sim 0{\cdot}5\mu$ in length, 300 Å wide, and 100 Å thick) and lay parallel to each other and perpendicular to the boundary.

INFLUENCE OF TEMPERING ON MICROSTUCTURE

The sequence of carbides formed after the different tempering treatments is shown diagrammatically in Fig. 13. This may be compared with Fig. 2 which shows the carbides formed after tempering the quenched steel. These two figures are almost identical except that in the normalized condition there is early local precipitation of Mo_2C and the persistence of this phase in the ferrite.

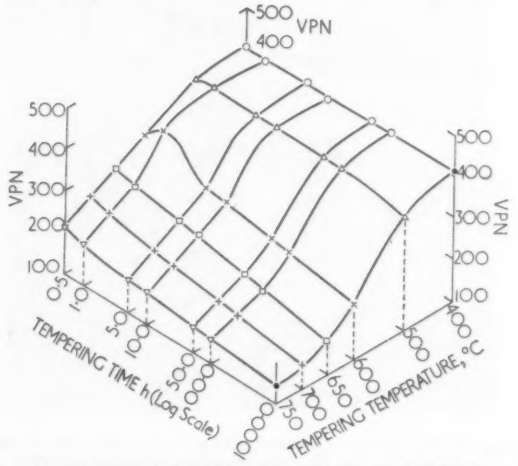


Fig. 11—Isometric diagram showing the variation in the VPN of the quenched steel on tempering

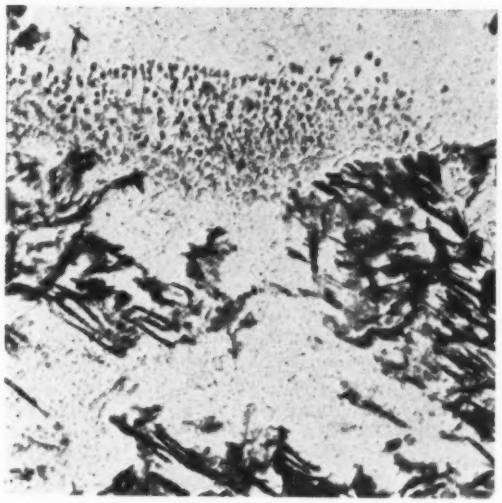


Fig. 12—Cr–Mo steel as-normalized, showing ferrite and bainite with some precipitation of Mo_2C at the interface; carbon extraction replica × 12 500

During tempering at 400° C, no change was observed in the ferrite regions but the bainitic carbides became gradually thicker. This change was accompanied by a gradual increase in the number of spots due to cementite in the diffraction patterns.

At 500° C, the bainitic carbides thickened more quickly than at 400° C whilst fringes of acicular Mo_2C were found at more of the ferrite/bainite boundaries. Small equiaxed particles were first precipitated very near the boundary, followed by the lath-like carbides growing from these nuclei into the ferrite as shown in Fig. 14. On tempering for long times, a diffuse zone-structure developed in both the pro-eutectoid ferrite and the ferrite regions of the bainite. This structure was very similar to that observed in the quenched condition after tempering for similar times.

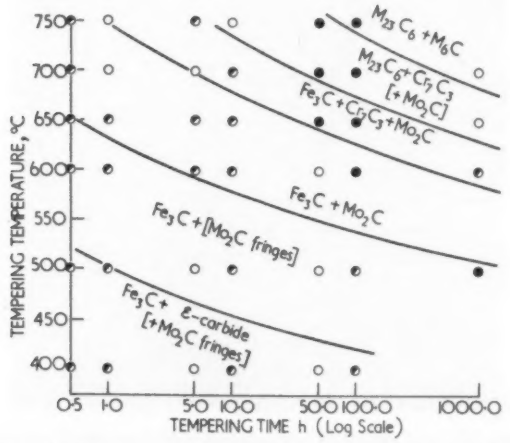


Fig. 13—Isothermal diagram showing the sequence of carbide formation on tempering the normalized steel

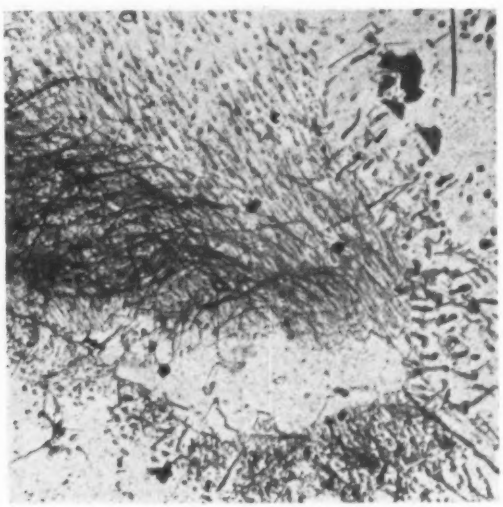


Fig. 14—Cr–Mo steel normalized and tempered for 5 h at 500° C showing Mo_2C needles growing from equiaxed nuclei; carbon extraction replica × 12 500

A diffuse zone structure developed in the ferritic matrix after only ½ h at 600° C (*see* Fig. 15) and, on further tempering, Mo_2C was precipitated, initially in the form of fine thread-like particles (*see* Fig. 16), which developed very quickly into well defined flattened needles. The bainitic carbides spheroidized markedly during tempering at this temperature. It appeared from some of the micrographs that the Mo_2C in the ferrite was located on some kind of sub-boundary network. The boundaries themselves were not discernible but the effect is clearly shown in Fig. 17.

At 650° C, the precipitation of Mo_2C was completed very much earlier than at 600° C and the cementite spheroidized more rapidly. Figure 18 shows Mo_2C precipitated at a ferrite/bainite boundary, the needles assuming a rather unusual configuration. After about 50 h, Cr_7C_3 was precipitated in regions near the bainite side of the bainite/ferrite boundary in characteristic stick-like form. Later $M_{23}C_6$ formed in the interior of the bainitic regions and recrystallization of the bainite began to take place. $M_{23}C_6$ was only observed in the bainitic regions and then only in those regions free from Cr_7C_3. Whilst the precipitation of the $M_{23}C_6$ carbides in the bainite was accompanied by solution of the adjacent Mo_2C needles, within the ferrite the Mo_2C was stable.

A similar sequence of charges occurred rather more rapidly on tempering at 700° C. Figure 19 shows the microstructure after tempering for 10 h. Some Cr_7C_3 can be seen at the ferrite/bainite boundary whilst the cementite is in an advanced state of spheroidization. The precipitation of $M_{23}C_6$ type carbide was observed to be accompanied by recrystallization of the bainite with new carbides appearing at the boundaries of the equiaxed ferrite grains. Eventually M_6C formed in both the ferrite and bainite regions and the Mo_2C in the ferrite at last began to dissolve.

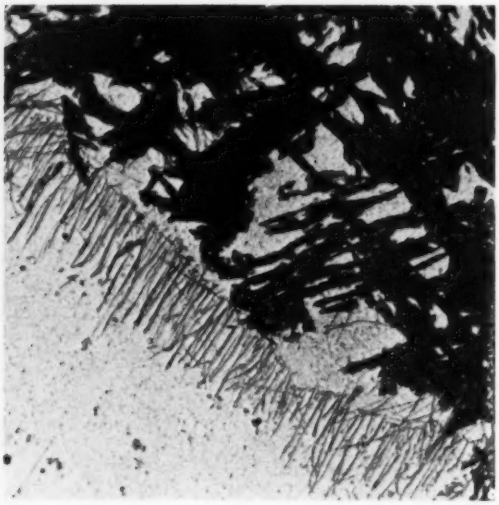


Fig. 15—Cr–Mo steel normalized and tempered for ½ h at 600° C showing massive cementite in the bainitic regions, acicular Mo_2C at the interface and ferrite; carbon extraction replica × 12 500

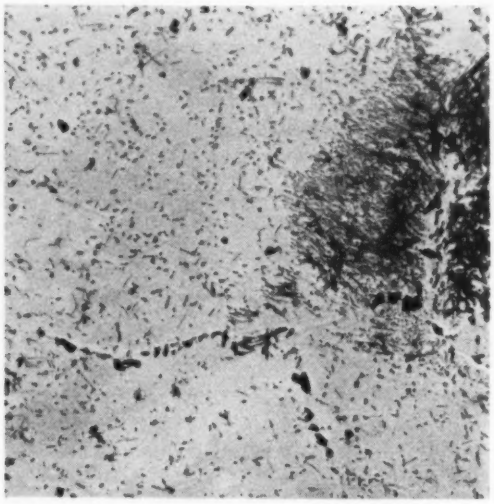


Fig. 17—Cr–Mo steel normalized and tempered for 100 h at 600° C showing precipitation of Mo_2C in the ferrite apparently on a sub-boundary network; carbon extraction replica × 3 000

At 750° C, the sequence of changes was the same as at 700° C. After 100 h there had been substantial grain growth in the bainite and much of the Cr_7C_3, $M_{23}C_6$, and Mo_2C had dissolved.

Influence of tempering upon chemical composition of extracts

The variation in the overall carbide composition during tempering is shown in Figs. 20–22. As in the quenched and tempered condition, there appeared to be replacement of Fe by Cr and Mn in the ε-carbide and cementite in the early stages of tempering. The Cr and Mn contents reached a maximum and then fell, but the effect was smaller than in the quenched and tempered steel. This is thought to be due to the fact that the appreciable amount of cementite present with ε-carbide in the bainite tends to mask the effect. The subsequent increase in the concentrations of Cr and Mn on further tempering was not quite as rapid as in the quenched and tempered steel. The overall concentration of Cr corresponding to the beginning of Cr_7C_3 formation was 12–16%, as against 16–18% in the quenched condition. When the only Mo carbides present were those at the boundaries, the overall Mo content was very small, but when general precipita-

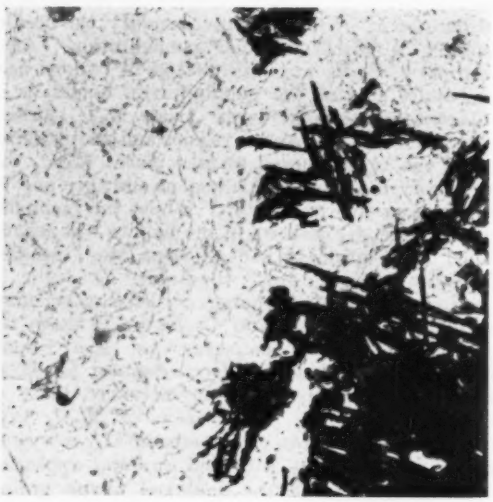


Fig. 16—Cr–Mo steel normalized and tempered for 5 h at 600° C showing precipitation of Mo_2C in the ferrite matrix; carbon extraction replica × 12 500

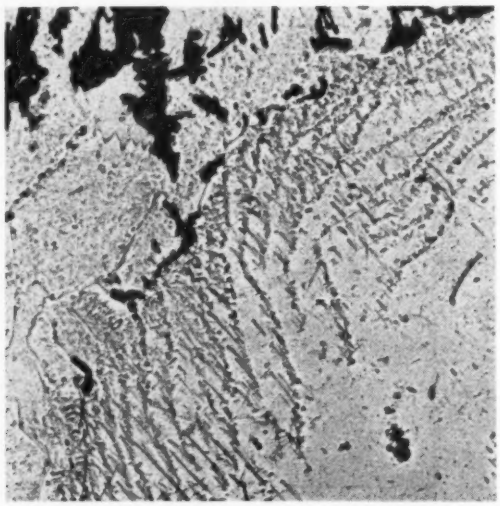


Fig. 18—Cr–Mo steel normalized and tempered for ½ h at 650° C showing Mo_2C precipitating in an unusual way at the ferrite/bainite interface; carbon extraction replica × 12 500

tion of Mo_2C occurred in the bainite and the ferrite, the Mo concentration increased and did not fall again even after 50 h at 750° C.

Influence of tempering on hardness

The effect of time and temperature upon the Vickers hardness is shown in the isometric diagram Fig. 23. In the as-normalized condition the hardness was 218. Very little change in hardness took place on tempering at 400° and 500° C, but there was slight secondary hardening caused by the precipitation of Mo_2C. Maximum hardness occurred slightly before Mo_2C needles were observed and when the microstructure still showed the diffuse ' zone ' structure. On further tempering, the hardness fell rapidly at first but the rate of softening decreased just before the formation of Cr_7C_3. When Cr_7C_3 began to form, the rate of softening again increased.

Microhardness tests were carried out in the ferritic and bainitic areas of the steel to determine their variation in hardness during tempering. The results are plotted isometrically in Figs. 24 and 25. The as-normalized bainite hardness was 336. The hardness changed very little on tempering at 400° and 500° C, but slight secondary hardening occurred. This was followed by a rapid fall in hardness and then a retardation in the rate of softening. During tempering the microhardness of the ferrite did not vary appreciably about a mean of 145. This result is rather surprising at first sight in view of the precipitation of Mo_2C which was observed to take place. The macrohardness of the composite bainite and ferrite was calculated by taking the weighted mean of the respective microhardness values for the ferrite and the bainite. The calculated values obtained agreed almost exactly with the experimentally determined Vickers hardness at a load of 50 kg. This is shown diagrammatically in Fig. 26.

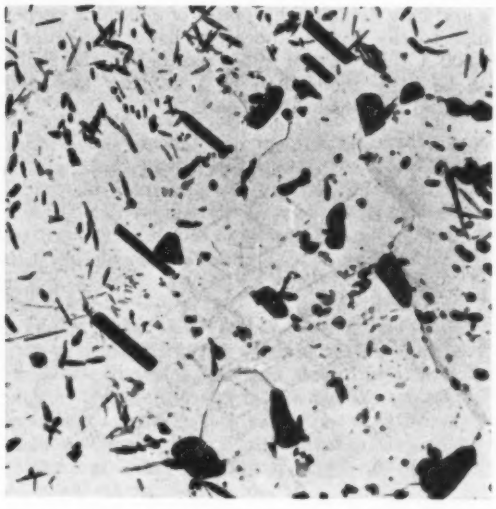


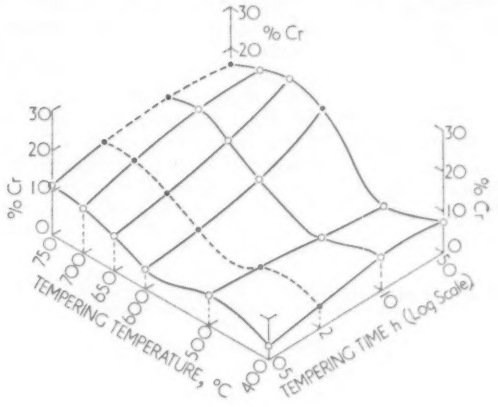
Fig. 19—Cr–Mo steel normalized and tempered for 10 h at 700° C, showing growth of the Mo_2C and spheroidization of the cementite in the bainite; carbon extraction replica × 12 500

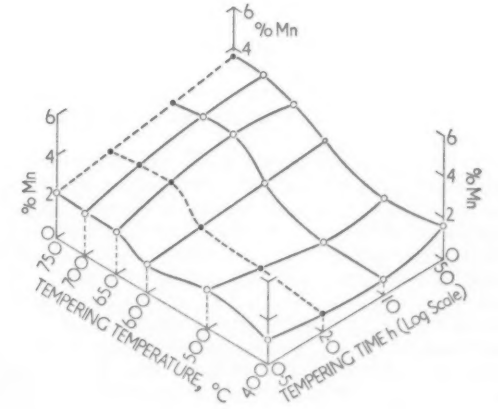
DISCUSSION OF RESULTS

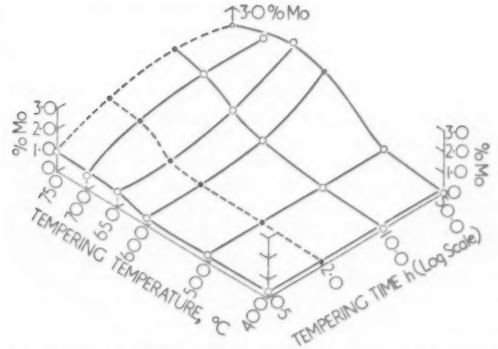
Nucleation and growth of the carbides

ε-Carbide

ε-Carbide was present in both the quenched and the normalized steels during the early stages of tempering.







Figs. 20–22—Isometric diagrams showing the apparent %Cr, Mn, Mo in the extracted carbides plotted against time and temperature of tempering for the normalized steel

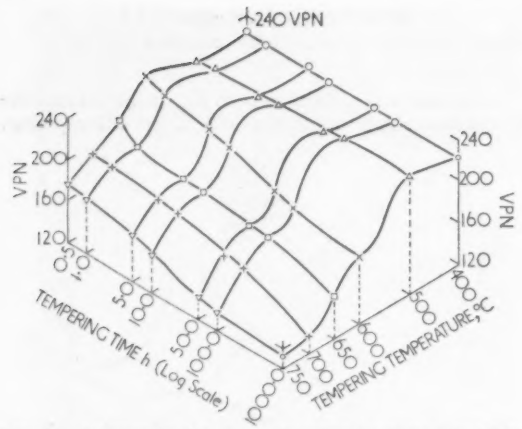


Fig. 23—Isometric diagram showing variation in VPN of the normalized steel on tempering

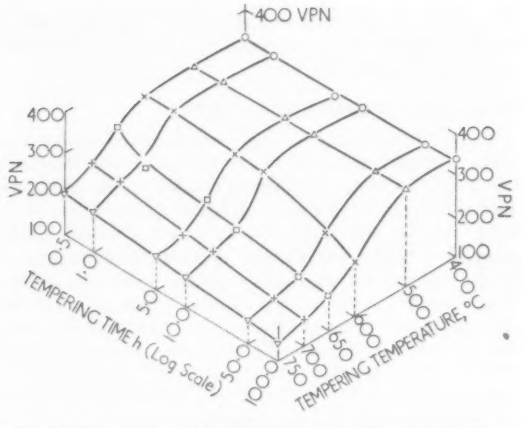


Fig. 25—Isometric diagram showing the variation in the microhardness of the bainite on tempering

The evidence for its presence is derived from three experimental sources:

(i) *Electron diffraction*—After tempering for only ½ h at 400° C, the electron diffraction patterns of the carbides are weak, diffuse, and contain only a few rings. These correspond to the more intense reflections of the ε-carbide pattern as given by Jack.[2] On further tempering, sharp spots due to cementite appear on the pattern, whilst the diffuse rings slowly disappear. However, since the ε-carbide rings are so diffuse and few in number, this evidence alone is regarded as inconclusive.

(ii) *X-ray fluorescence analysis*—The X-ray fluorescence analysis results are difficult to explain except on the assumption that one carbide is transforming to another in which the solubility of Cr and Mn is less. The carbide present after the fall in alloy content is certainly cementite, thus the inference is that ε-carbide is present before the fall in alloy content occurs. Further evidence is provided by the increase in the total amount of carbide which accompanies the fall in alloy content. Previous work[3, 4] has shown that ε-carbide may be in metastable equilibrium with a much larger amount of carbon in solution than cementite, therefore during the transformation of ε-carbide to cementite the volume fraction of the carbides should increase.

(iii) *Electron microscopy*—The changes in the precipitate morphology which take place on tempering at 400° and 500° C are in close agreement with those reported by Crafts and Lamont[5, 6] in their investigation of the transformation of ε-carbide to cementite using the electron microscope. Thus the present work confirms the results of Smith and Nutting[7] that ε-carbide may occur in tempered low-alloy steels containing only about 0·2%C. It was observed both by Smith and Nutting and by the present authors that the outlines of the ε-carbide particles were ragged in appearance. This is believed to be due to localized attack by the etchant at the grain-boundaries. Following the ideas of Kinzel,[8] it is thought that this form of attack may occur when some coherency exists between the precipitate and the matrix. This 'gouging' attack increases the apparent size of the precipitate and may explain why plates of ε-carbide perpendicular to the replica appear to be of the order of 400–500 Å thick, whilst those parallel to the replica must be little more than 50 Å as they are easily penetrated by the electron beam.

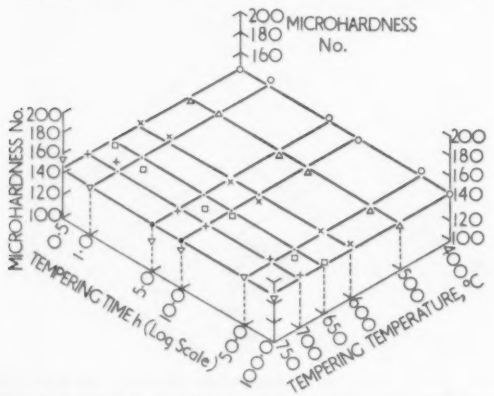


Fig. 24—Isometric diagram showing the variation in the microhardness of the ferrite on tempering

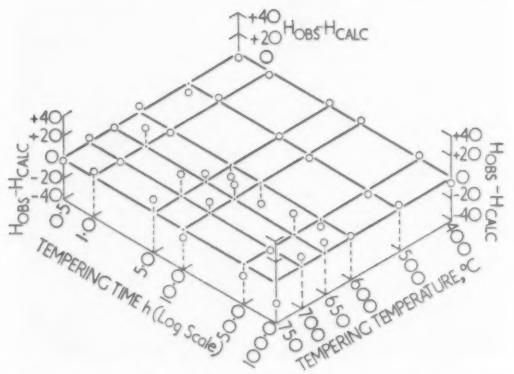


Fig. 26—Isometric diagram showing the difference between VPN of the normalized and tempered steel and the weighted mean of the microhardness of the ferrite and bainite

Cementite

During the formation of cementite, there is no evidence of further nucleation in the matrix and no change in particle shape except for the slow thickening of the plates. This suggests that the ε-carbide → cementite transformation takes place *in situ* perhaps in a manner similar to that suggested by Jack.[2] It is interesting to note that, during subsequent tempering, the Widmanstätten precipitates within the grains are very resistant to spheroidization which might indicate that they are partially coherent with the matrix. This might be expected if Jack's mechanism for the transformation is correct, since good fit should still obtain in at least one direction between the cementite and the matrix.

Cr_7C_3

The experimental evidence indicates that Cr_7C_3 nucleates in the vicinity of cementite in these steels, probably at the interface between the carbide and the ferrite. The chromium content of cementite increased before the formation of Cr_7C_3, a result obtained also by other workers.[9, 10] As Kuo[9, 11] has pointed out, separate nucleation would be unlikely in a matrix already impoverished in Cr. Further, no metallographic evidence of precipitation by separate nucleation was observed, whilst the carbides grew rapidly in the early stages. Perhaps the most important evidence, however, was that nucleation of Cr_7C_3 in the normalized steel only took place in the bainitic regions, although the ferrite might be expected to be the richer in Cr; i.e. Cr_7C_3 was only found where cementite had previously existed.

$M_{23}C_6$

$M_{23}C_6$ was formed throughout the matrix of the quenched and tempered steel but only in the bainite regions of the normalized and tempered steel. It formed in the interior of the bainite grains in the regions containing cementite and Mo_2C and grew at the expense of these carbides. It was never observed in the vicinity of Cr_7C_3 colonies, nor did the Cr_7C_3 appear to dissolve after $M_{23}C_6$ had been formed. Thus, the $M_{23}C_6$ carbide is probably based on Fe and Mo, and is possibly of the $Fe_{21}Mo_2C_6$ type. If it were chromium-rich, it would be expected to grow at the expense of Cr_7C_3 as was observed by Shaw and Quarrell[10] in high-chromium steels. It is thought that nucleation of $M_{23}C_6$ takes place in the vicinity of the cementite particles. This would explain the rapid growth of the $M_{23}C_6$ in the early stages.

M_6C

The M_6C particles form at grain boundaries and grow very rapidly at the expense of all the surrounding carbides. In the normalized steel, nucleation occurs both in the ferrite and in the bainite regions. It is thought that nucleation takes place at the existing carbide particles, which might explain why only a few particles are formed and why they grow so rapidly in the early stages.

Mo_2C

Mo_2C was precipitated after tempering both the quenched and the normalized steels, in each case by separate nucleation in the ferrite. This is in agreement with the work of Kuo,[11] Smith and Nutting,[7] and Honeycombe and Seal.[12] The changes in microstructure observed during the nucleation process are very similar to those observed in the early stages of the precipitation of V_4C_3.[7] As in the case of V_4C_3, Mo_2C is believed to be completely coherent when first formed. This corresponds to the condition for maximum hardness. Later, when the Mo_2C grows into well defined flattened needles, it is believed to have lost some of its coherency with the matrix. Habraken[13, 14] who has made an extensive investigation of steels of this type, believes that the needles follow the <110> directions on {100} planes of the matrix. The metallographic evidence obtained in the present investigation appears to be in accordance with this view.

Formation of fringes of Mo_2C

The formation of fringes of acicular carbides at the austenite grain boundaries has been observed previously by Schrader[15] and Fisher.[16] They have been identified as Mo_2C in this investigation by selected area electron diffraction.

Their mode of formation during normalizing or after normalizing and tempering may be explained as follows. When the pro-eutectoid ferrite is first formed on cooling from the austenite range, it becomes enriched with alloying elements such as Cr and Mo, whilst the remaining austenite becomes enriched with carbon. At the boundaries between the ferrite and austenite, the local concentration of both Mo and C is high and precipitation of Mo_2C may occur at the boundary. The subsequent rate and direction of growth will be controlled by the relative rates of supply of Mo and C. Since the rate of diffusion of C is much greater than that of Mo the direction of growth will be such as to minimize the diffusion path of the Mo. Therefore the precipitate grows into the Mo-rich ferrite.

If the cooling rate is sufficiently rapid, the formation of Mo_2C fringes during normalizing is suppressed. However, if some austenite is retained during cooling then on tempering Mo_2C fringes may be formed once again at the bainite/ferrite interface.

Electron diffraction indicates that the Mo_2C particles are similarly orientated in all respects and are thought to be growing on the favourable growth system in the ferrite most nearly perpendicular to the interface between the bainite and the ferrite.

Influence of carbide particles on hardness

Although precipitation occurs in the ferritic regions of the normalized steel during tempering, the hardness does not appear to alter appreciably.

The mean particle spacing of the carbides within the pro-eutectoid ferrite of the normalized and tempered steel was determined after initial precipitation was complete. Similar determinations were made for the carbides formed in the quenched and tempered steel. It was found that, at the same mean carbide spacing, the hardness of the normalized and tempered steel was 100 units lower than the hardness of the quenched and tempered steel. It has been suggested by Hyam and Nutting[17] that unless coherency hardening is occurring, the carbide particles have no direct effect

on the hardness except insofar as they control the ferrite grain size. The present investigation seems to afford support to these suggestions. It has been shown by the present authors[18] that in the case of vanadium steels the coherent V_4C_3 particles are surrounded by strain fields in the same way as the GP zones in the Al–4%Cu.[19] While the particles are fully coherent with the matrix, the strain fields may be quite extensive and it is the interaction of these fields with dislocations which is thought to cause the hardening.[12] When the particles lose coherency, the extent of the strain fields is reduced. Thus the direct effect of non-coherent particles on the hardness is probably very small compared with that of coherent particles. Nevertheless the non-coherent particles might be expected to have an appreciable effect on the hardness if they could act as efficient obstacles to dislocations during deformation. However, it is thought that such an effect would never be very large in the case of an iron matrix. Thomas and Nutting[20] have shown that even in the face-centred cubic aluminium-based alloys slip can pass round non-coherent precipitates and it is thought that in the case of body-centred cubic iron, with its many possible slip systems, this process would be much easier and occur in a manner analogous to that suggested by Hirsch.[21]

Influence of microstructure on creep properties

The 2¼%Cr–1%Mo steel has better creep properties at 500° C after normalizing and tempering than after quenching and tempering. At first sight this seems a little surprising since after identical tempering treatments the quenched and tempered steel always has the greater hardness. However, as Cottrell[22] has stated, the most important factor in long time creep is microstructural stability. One of the best ways of obtaining this in low-alloy ferritic steels is to form a stable dispersion of an alloy carbide which is both resistant to growth and to any further phase-change. In normalized and tempered steel the carbide Mo_2C forms a dispersion which persists until the formation of M_6C but in the quenched and tempered steel the Mo_2C disappears much earlier. This is undoubtedly because $M_{23}C_6$ (which forms at the expense of Mo_2C) occurs only in the bainitic regions of the normalized steel but throughout the microstructure of the quenched steel.

The exact role of the Mo_2C in promoting creep resistance is not clear. Clearly, in the amounts observed, the precipitate has little effect on the hardness at room temperature but possibly it might have a greater relative effect on the passage of dislocations at higher temperatures, which would result in the stiffening of the grains and thus make deformation more difficult.

The effect of precipitates on grain boundary sliding is not clear. It is possible that the precipitate and in particular the fringes of Mo_2C at the boundaries retard sliding. However, it is thought more likely that the principal effect of the precipitate is to prevent boundary and sub-boundary migration. It is well known that in pure metals relaxation of creep stresses can occur by boundary migration. This may be prevented by the presence of a precipitate. It is also known that cell formation occurs during creep and that these cells or subgrains are regions of perfect lattice separated by dislocation networks. There is some evidence to suggest that Mo_2C is located at these cell boundaries where it probably has a stabilizing influence.

SUMMARY AND CONCLUSIONS

(1) On tempering the quenched steel, the sequence of changes in the carbide structures was:

$$\epsilon\text{-carbide} \rightarrow \text{cementite} \rightarrow \begin{cases} \text{cementite} \\ + \\ Mo_2C \end{cases} \rightarrow \begin{cases} \rightarrow Cr_7C \\ \rightarrow M_{23}C_6 \rightarrow M_6C \end{cases}$$

(2) On tempering the normalized steel, the sequence of changes in the carbide structures was:

(i) *Bainite*

$$\begin{cases} \epsilon\text{-carbide} \\ + \\ \text{cementite} \end{cases} \rightarrow \text{cementite} \rightarrow \begin{cases} \text{cementite} \\ + \\ Mo_2C \end{cases} \rightarrow \begin{cases} \rightarrow Cr_7C_3 \\ \rightarrow M_{23}C_6 \rightarrow M_6C \end{cases}$$

(*ii*) *Ferrite*

$$Mo_2C \rightarrow M_6C$$

(3) ϵ-carbide occurs in both the normalized and in the quenched conditions after tempering. It is more stable than in plain carbon steels and is not completely transformed to cementite until after 5 h at 500° C. Replacement of Fe by Cr and Mn occurs but Mo does not appear to dissolve appreciably.

(4) ϵ-carbide transforms to cementite in situ, the carbide plates thickening and becoming sharper in outline. Replacement of Fe in cementite by Cr and Mn can occur and the solubility of Cr increases with temperature.

(5) Cr_7C_3 forms by nucleation either within the cementite particles or at the cementite/ferrite interface.

(6) $M_{23}C_6$ grows at the expense of both cementite and Mo_2C. It is believed to nucleate either within the cementite particles or at the ferrite/cementite interface.

(7) M_6C may form at the expense of Mo_2C, $M_{23}C_6$ or Cr_7C_3. It is probably nucleated at the interfaces between existing carbides and the matrix.

(8) Mo_2C precipitates by separate nucleation on tempering both quenched and normalized steel. Diffuse zones appear in the matrix followed by the growth of well defined flattened needles from these zones.

(9) Retained austenite occurs in the quenched steel. It is apparently stabilized by one or more of the alloying elements present since it only decomposes on tempering at 500° C. The austenite decomposes with the precipitation of large carbides of the cementite type, closely resembling those observed in bainite.

(10) During slow cooling from the austenite range, Mo dissolves preferentially in the pro-eutectoid ferrite whilst C becomes concentrated in the remaining austenite. This leads to the precipitation of Mo_2C in characteristic fringes at some of the bainite/ferrite boundaries.

(11) Recrystallization accompanies the formation of $M_{23}C_6$ both in the quenched steel and in the bainitic regions of the normalized steel. It probably occurs because of the great increase in mean particle spacing

which accompanies the formation of this carbide from cementite and Mo_2C.

(12) The hardness of the bainite/ferrite aggregate in the normalized steel is approximately the weighted mean of the respective microhardness of the bainite and the ferrite.

(13) Precipitation of Mo_2C in the quantities observed has no apparent effect on the hardness of the ferritic regions of the normalized steel. The density of precipitation is probably insufficient to affect the hardness appreciably, but in any case the hardening effect of partially coherent particles is probably small.

(14) The steel has better creep resistance at 500° C in the normalized and tempered condition because of superior microstructural stability. This is undoubtedly connected with the persistence of Mo_2C in the normalized steel, but the exact role of this precipitate is not clear.

ACKNOWLEDGMENTS

The authors would like to thank the members of the BISRA High-temperature Steels Committee for their constant encouragement and advice and in particular Mr P. Whittaker of Stewarts and Lloyds Ltd for the provision of steel for this investigation. The X-ray fluorescence analyses were carried out at Eindhoven on the Philips X-ray fluorescence analysis instrument in the laboratories of that company and the authors are grateful to Dr de Vries for his advice and assistance. They are also grateful to Dr J. F. Radavich (Purdue University) for helpful suggestions regarding X-ray techniques.

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APPENDIX I

Scheme of experimental investigation

The steel, of percentage analysis

C	Si	S	P	Mn
0·15	0·18	0·012	0·018	0·50
Cr	**Mo**	**Cu**	**Ni**	**Sn**
2·12	0·94	0·077	0·165	0·009

was received in the form of a short segment of steam pipe in the as-forged condition. It was cut into specimens 1 in. × ½ in. × ½ in., the longest dimension being parallel to the pipe axis. Austenitizing was carried out by heating for 2 h at 960° C after which half the specimens were cooled at 300° C per hour to room temperature and the remainder were quenched in oil. The subsequent tempering treatments were carried out in a lead-bath at 400°, 500°, and 600° C and in an electric air furnace at 650°, 700°, and 750° C. In each case the temperature was maintained constant to within ± 2° C. All the specimens were water-quenched after tempering.

After heat treatment, the specimens were sectioned and the exposed surfaces prepared for hardness testing and metallographic examination by abrading on emery papers, electropolishing, and etching in nital. The VPN of each specimen was determined from measurements made on 5 indentations, using a load of 50 kg. In addition, microhardness determinations were carried out on the normalized and tempered specimen using a Reichert microscope with microhardness attachment and loads of 20 and 50 g. Carbon extraction replicas were prepared using the method developed by Smith and Nutting.[1]

Using the direct carbon extraction replica as the basic extraction technique, four methods were used to investigate the extracted carbides. In addition to the techniques of electron microscopy and electron diffraction, two new preparation methods were developed which enabled X-ray diffraction and X-ray fluorescence analysis to be used with the extraction replicas.

To mount the material for X-ray diffraction, a fine glass fibre was mounted in a holder and used to pick up pieces of extraction replica floating on a water surface. The collected fragments were dried in a current of warm air, then the fibre was dipped into a dilute plastic solution and again dried. The plastic fulfilled the dual function of cementing the pieces of replica in position and preventing the carbides from oxidizing. The X-ray diffraction photographs were taken in a 5·7 cm Philips Debye–Scherrer-type camera using Cu radiation and Ni foil filter between specimen and film. Exposures of about 2 h at 20 mA were found to give a satisfactory result for amounts of carbide estimated at between 2 and 5 μg.

For X-ray fluorescence analysis, the specimens to be analysed were irradiated with X-rays of high energy, causing the elements present to emit secondary radiation which was passed through a collimating system to an analysing crystal at the centre of a goniometer system. The intensity of the diffracted rays of any particular wavelength could be measured by a G–M counter, which could rotate about the same axis of rotation as the crystal but with twice its angular velocity, thus satisfying the condition for Bragg reflection. The specimens for analysis were prepared by mounting pieces of carbon extraction replica on 'Melinex' films, the procedure being as follows. A little alcohol was added to the water in which the pieces of replica were dispersed. This lowered the surface tension sufficiently to enable filtration to take place through fine-mesh stainless-steel gauze. The films were thoroughly washed with alcohol and collected into a small depression near the edge of the gauze, using a glass rod as a 'policeman.' The films were transferred to the centre of a piece of 'Melinex' about 3 in. square and dispersed over a circular region about ½ in. dia. After the evaporation of the alcohol, the films continued to adhere to the plastic and the 'Melinex' could be folded and stored in an envelope.

Representative samples of the carbides were analysed qualitatively by allowing the apparatus to scan automatically over a range of values of 2θ between 20° and 70°. The results were recorded on a chart by a pen-recorder. The goniometer was then set for the counter to receive radiation from each of the elements in turn for a counting time of 2 min. The apparent concentration of each element in the extracted carbides was then calculated as follows:

$$\text{At-}\%\ E_1 = \frac{C_{E_1}}{\sum_{x=1}^{x=N} C_{E_x}} \times 100$$

where C_{E_1} = number of counts from element E_1 in 2 min.

$\sum_{x=1}^{x=N} C_{E_x}$ = total number of counts during a 2-min period from all the elements recorded

Whilst the analyses obtained are not absolute, the results indicate definite trends in the chemical composition of the particles extracted in the replicas. The counts

are subject to a random error due to fluctuations in the rate of counting, the probable error in any count of magnitude being $C_E^{\frac{1}{2}}$

If $\sum C_E \simeq 10^4$ counts/min

and if two counts are taken, then for an apparent concentration of 1% the probable error is about 30% of this, i.e. $\pm 0{\cdot}3\%$.

For an apparent concentration of 10%, the probable error is about 5% of this, i.e. $\pm 0{\cdot}5\%$. In most of the specimens $\sum C_E \geqslant 10^4$ counts/min so that this can be regarded as a limiting error.

It is very difficult to calculate the systematic errors, but it would be possible to eliminate them by using standard specimens for calibration purposes. In the results presented the trends in composition are correct, but the apparent percentages quoted do not correspond with the true percentages.

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Letter to the Editor

Austenite transformation and the Greninger–Troiano technique

THE TRANSFORMATION of austenite to martensite is generally studied by the Greninger–Troiano (G–T) technique, where the proportion of martensite formed from austenite on quenching to temperatures in the M_s–M_f range is microscopically estimated. The direct quench is followed by a short isothermal hold at the temperature of quench, a reheat to a temperature just above M_s which allows the martensites already formed to be tempered, an isothermal hold at this latter temperature, and a final quench. Minor modifications seem necessary, however, to the results obtained by this technique, as the effects on the transformation of austenite of factors inherent in the technique itself and of the isothermally formed martensite and bainite do not appear to have been fully taken into account. These are now briefly considered, errors arising from microscopic estimation being ignored.

The most important factors affecting the austenite–martensite transformation are the thermal stabilization induced above the M_s temperature due either to a lower heat abstracting capacity of the quenching bath or to a temperature of quench closer to the M_s, the interruption of the quench, the martensite formed up to that stage, and the isothermal hold before the final quench. The isothermal formation of martensite and/or bainite at certain stages of the G–T cycle also alters the final distribution of the micro-constituents.

An estimate of the effect on the retained austenite of the thermal stabilities induced above the M_s temperature by quenching in different baths as well as to different temperatures below the M_s has been made from data derived from the experimental results of Lueg and Pomp.[1] For an M_s temperature of 200° C and a bath temperature of 180° C, the retained austenite differs by about 3% between metal and salt bath quenching. Quenching to 180° and 70° C respectively in a metal bath, however, additionally changes the amount of retained austenite by about $7{\cdot}5\%$. This is because of the higher stability of the austenite at quenching temperatures close to the M_s.

On interrupting the quench and holding isothermally, a partial transformation of the retained austenite to martensite occurs, while the stability of the residual austenite is increased; from available experimental data, the effect of the former on the final retained austenite content is estimated at about 6–8% and that of the latter at 2–3%. Austenite will also transform to martensite as its effective M_s temperature will be raised following carbide precipitation.[2] The extent of such transformation will depend, obviously, upon the time of hold and may well be negligible if holding times are selected only to attain thermal equilibrium.

An examination of Morgan and Ko's results[3] for the thermal stability induced in austenite, isothermally held at temperatures above M_s, shows that such induced stability drops steeply with the approach of the holding temperature to the M_s value for all reasonable holding times. As in the G–T technique, the temperature of the second isothermal hold is generally only slightly above M_s, it is considered that this effect is too small to be taken into account. The final quenching would appear to need no special consideration.

We thank Dr B. R. Nijhawan, Director of the National Metallurgical Laboratory, for kind discussion and permission to publish this letter.

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15th May, 1959
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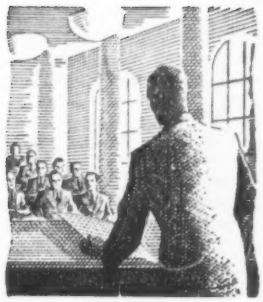
DISCUSSION ON PAPERS

Discussion at Meetings · Written Contributions



Discussion at the Autumn Meeting, 1959

A report of the AUTUMN GENERAL MEETING was given in the March, 1959 issue of the *Journal*. Discussions at some of the technical sessions are given below; further discussions will be published later in the *Journal*.

The discussion on continuous casting was preceded by some introductory remarks by Mr G. N. F. WINGATE concerning the papers discussed at Charleroi during the Journées Internationales de Sidérurgie. These are also given below.

Discussion on continuous casting

Part 1 Morning session

The following papers were introduced by Mr Wingate (dates of publication in the *Journal* are in parentheses):

"Continuous casting on the BISRA experimental plant" by *G. Fenton* and *J. Pearson* (1958, vol. 189, pp. 160–167)

"Continuous casting at the Centre National de Recherches Metallurgiques, Section du Hainaut, Charleroi, Belgium" by *J. Zaeytydt* (1958, vol. 190, pp. 165–168)

"Control of flow of liquid metal during continuous casting" by *W. Siegfried* and *B. Broniewski* (1958, vol. 190, pp. 162–165)

"Continuous casting of steel in France" by *P. Thomas* (1958, vol. 190, pp. 112–113)

"Continuous casting of steel in Western Germany" by *K. G. Speith* and *A. Bungeroth* (1958, vol. 190, pp. 158–161)

"New experience in the continuous casting of steel" by *H. Krainer* and *B. Tarmann* (1958, vol. 190, pp. 105–111)

"Continuous casting at Atlas Steels Ltd, Welland, Ontario, Canada" by *G. C. Olson* (1958, vol. 190, pp. 40–50)

Mr **Wingate** said: At Charleroi, many of us felt for the first time that continuous casting had become an established industrial process. This leads me to try to set the keynote by reviewing briefly the more important industrial applications of continuous casting in the main steel-producing countries of the world and, in doing so, I shall draw freely on some of the papers presented at Charleroi.

INDUSTRIAL APPLICATIONS: REVIEW

There are various qualities of steel being cast in the plants currently in production, for continuous casting is no longer regarded as a process suitable only for highly alloyed steels made in small quantities. The casting of carbon steels and low alloy varieties is of primary interest in most of the installations, and it seems reasonable to conclude that, where local conditions allow, continuous casting can be applied to the production of almost all types of steel.

USSR

There already exist in the USSR at least five continuous-casting plants, and others are being built or planned. One plant has produced over 100 000 tons of good slabs, and is operating 7 days a week, 3 shifts a day, and producing 4 000 tons/month.

The trend in the plants under construction is towards the casting of larger slab sections from ladles of larger capacity and to the operation of four-strand, and later eight-strand, machines. At the new Stalino plant, it is hoped to cast rimming steel from ladles of 140 tons capacity on a four-strand machine with 1 200 × 200 mm (48 in. × 8 in.) moulds. Under consideration is an eight-strand machine suitable for operation with 250-ton ladles and having an annual production of 2 000 000 tons. These are vast tonnage projects.

Japan

Sumitomo Metal Industries have been casting billets up to 170 mm (6¾ in.) square, and 200 mm × 75 mm (8 in. × 3 in.) slabs on a machine supplied with 10-ton ladles. Production has been up to 800 tons/month, and a second machine is under construction to operate with a ladle capacity of 40 tons.

Another Japanese company is proposing to install a machine to cast 1 200 mm × 125 mm (48 in. × 5 in.) slabs initially and, later, large-diameter rounds in two strands. This plant will be supplied with arc-furnace metal in 40-ton ladles.

Germany

The well known plant at Mannesmann-Hüttenwerke is a four-strand machine, taking 35 tons of metal and casting up to 300-mm (11¾-in.) dia. rounds and a variety of other sections, on a semi-commercial basis with an output of about 800 tons/month, mainly of low-alloy electric steel for tubes.

This year, production began on a new machine at the Benteler works in Neuhaus/Paderborn. Supplied with arc-furnace metal in 18-ton ladles the machine single-strands 120-mm (4¾-in.) square billets but is designed to

cast billets up to 160 mm (6¼ in.) square and 300 mm × 75 mm (12 in. × 3 in.) slabs. The product from this plant is also being used for tube production.

France

The Jacob Holtzer plant at Unieux, casting small oval sections of high-quality special high-alloy steels, has been in regular production for some years.

At the Cail works in Denain a four-strand machine is in regular production, casting 240-mm (9½ in.) square blooms. The machine is supplied with arc-furnace metal in a 35-ton ladle. Over 100 000 tons of good castings have so far been made at this plant.

The plant at Allevard uses a 12-ton ladle and regularly casts 75-mm (3-in.) square and 140 mm × 90 mm (5 in. × 3½ in.) rectangular sections, mainly of Si–Mn spring steel. Allevard are erecting a second machine and, thereafter, will 'dispense with blooming.'

Austria

The Böhler plant at Kapfenburg is well known. Supplied with 7–8 tons of electrically melted steel, it produces a wide range of sections.

At Breitenfeld there is a twin-strand machine, casting 110- and 130-mm (4½- and 5-in.) dia. rounds, some rectangular sections and 275 mm × 65 mm (10¾ in. × 2½ in.) slab sections. This is reported in Dr Krainer's paper as handling the greater part of the steel produced at this works.

At Ternitz, Schoeller-Bleckmann are operating a plant supplied with open-hearth and arc-furnace metal in 18- and 9-ton ladles. This single-strand machine has been casting 100-mm (4-in.) square billets of structural steel but, later, it is planned to cast 500 mm × 180 mm (19½ in. × 7 in.) slab sections in 18–8 and titanium-stabilized steel.

Italy

At Terni, an eight-strand, or rather two four-strand machines have recently been installed to cast O H metal from 65-ton furnaces and arc-furnace metal from 35-ton furnaces. The casting time with a 60-ton ladle is expected to be about 1 h, and the sections being cast are 110- and 140-mm (4½- and 5½-in.) squares. The total output should reach about 200 000 tons/year.

At Udine the new SAFAU machine is fed with ladles up to 30 tons capacity; the machine is designed to cast 200–250-mm (8–10-in.) square billets and, whilst casting rates of 20 tons/h have been achieved on a 240-mm (9½-in.) square section, the aim is to raise the production rate to 30 tons/h.

Sweden and Norway

The plant at Upsala is of small capacity, casting 125-mm (5-in.) dia. billets from 1½-ton charges, whilst the plant at Nyby Bruk is supplied with 5-ton and 15-ton charges to single strand mainly 330 mm × 55 mm (13 in. × 2¼ in.) slab sections.

Early in 1959 a machine will begin operation at Stavanger in Norway, with 8-ton and 12-ton ladles of arc-furnace metal. At this plant 165-mm (6½-in.) square billets will be cast and the expected annual production is 10 000 tons of carbon steel.

Switzerland

Under construction for the von Moos works near Lucerne is a single-strand machine of 15-ton capacity for the production of 85–150 mm (3¼–6 in.) square billets for conversion to wire, etc.

Spain

At the new Del Besos works near Barcelona a machine is now being erected to cast from a 15-ton ladle 50–100 mm (2–4 in.) squares and 200 mm × 60 mm (8 in. × 2⅜ in.) slabs of plain-carbon arc-furnace steel.

Canada

The only plant in operation is that at Atlas Steel, where 6¼-in. square billets and 24 × 5½ in. slabs have been cast. Current interest is in the production of thin slabs, roughly 17½ in. × 2½ in., for direct feeding to the Sendzimir mill. This plant is supplied by arc-furnace metal and, for the most part, casts stainless steel. Until recently the machine at Atlas was supplied with a 35-ton ladle, but recent work has included the disposal of metal from 50-ton ladles. The machine is either a single-strand or twin-strand unit, depending on the size of section cast.

USA

Although there is fairly widespread interest in one or two small steelworks in the United States, at present there are no plants producing continuous-cast material on a regular basis.

UK

In the UK, apart from the BISRA plant, there are only two plants at present working in the continuous-casting field, one is at the Low Moor Alloy Steel Works near Bradford, where squares and rounds up to 5½ in. and slabs up to 14 × 3 in., mainly in alloy steels, are produced.

The other is at Barrow Steel Works. Here, ladles of 7½-tons capacity are cast with some regularity. The plant is a high-speed unit designed for the casting of small sections, 2–4-in. square billets and slab sections up to 12 in. × 3 in. The machine is a twin-strand one, equipped with automatic control, and from a technical angle represents a fairly high degree of development.

INDUSTRIAL APPLICATIONS: THE FUTURE

From this brief review of world developments in the continuous-casting field, it should be clear that there are wide and far reaching possibilities for the application of continuous casting as a principal production process. To those who may be interested in trying for the first time a new basic process of this type, the advantages to be gained are of primary importance and, above all, the economic aspects need careful consideration.

ECONOMICS

At present no figures are available for any plant working with its full production through continuous-casting machines and the economy of such a works has therefore to be gauged rather from estimates of the technical advantages of the process.

General experience to date with continuous casting shows that the main advantage of high overall average yields can be obtained. Equally important, the quality of the product produced is good and, in some cases, better than that from the orthodox ingot process. Operational advantages come because the product generally requires less dressing and the high yields allow considerable reduction in scrap recycling and in the fuel consumption required to produce each final ton of steel.

These advantages, higher yield, good quality, lower fuel consumption, taken alone appear sufficiently great to induce some steelmakers to substitute continuous casting for conventional casting, particularly where ingots of small cross-section were being cast formerly by orthodox means, e.g. Cail and Terni, plants mainly casting carbon and low-alloy steels. If the casting machine conversion costs are greater than the orthodox pit conversion costs, then some of the savings from the improved yields must be nullified. Here we have low-priced steels where the cash returns from an improved

yield are much less than with high-alloy steels. Can it be that under certain circumstances, in these cases four- or eight-stranding, conversion by continuous casting is no more costly than by conventional pit casting? Should we think of a continuous-casting machine, not a very expensive piece of equipment, as in some ways the equivalent of a conventional pit-side installation with its teeming stage, its bogies or ladle casting cars, its moulds and bottom plates? There is a lot of capital tied up in a pit side. Is the manning for the machine necessarily any greater than that for the conventional pit? Is there a minimum throughput? Terni think that in their case this minimum is 5 000 tons/month.

This is a challenging idea to some of us who have hitherto thought that the economics of continuous casting must depend mainly on yield savings on highly priced steels or on eliminating intermediate forming processes or on providing a marginal increase in production which cannot be catered for by an existing rolling mill.

It remains true that one of the chief attractions of continuous casting is that in suitable cases it can eliminate an appreciable amount of cogging and intermediate rolling. Continuous-casting plant is relatively inexpensive so that one can avoid heavy capital expenditure on blooming mills, only justifiable when large tonnages are involved.

Small tonnage applications

Apart from capital cost, mild-steel billets are by orthodox means difficult to produce economically, except in very large tonnages. So continuous casting has an appeal to the small isolated steel producer, or would-be producer, with limited raw materials and limited markets for his finished product. With a production target of perhaps 50 000–100 000 tons/year he will probably incline to small arc furnaces and these, coupled with casting machines to cast the smallest re-rolling sections, may enable him to produce billets as economically as he could buy them elsewhere. As indicated already, the casting machines will be a minor item of expenditure, the major capital cost being on furnaces, buildings, cranes, and other services. The capital cost for such a complete new plant might be £20 per ton of annual production.

Large tonnage applications

In regard to the casting of larger sections I think it is fairly obvious, particularly from the Soviet developments, that major advantages and savings will result from multi-stranding if the overall production rate can be increased to match that of large-capacity O H furnaces. The production rate for larger sections increases roughly in proportion to the cross-sectional area of the section, whilst multi-stranding allows the casting period for disposal of a full furnace charge to be brought within workable limits, something of the order of 1 h.

Large tonnage installations must each be considered individually. As with other plants, because of the greater tonnage throughput, overhead costs are much reduced and it is worthwhile, therefore, giving particular attention at the construction stage to special arrangements most suited to the product concerned.

Of particular interest in this country would be plants for the production of large slabs suitable for conversion in continuous rolling mills to broad sheet. Generally, the large slab machines installed to date cast relatively thick slabs of dimensions similar to those produced by orthodox slabbing mills. However, the production of thin slabs of considerable width would appear to be a straightforward development and would be more advantageous if allied with the method of horizontal discharge by bending. No loss of tonnage throughput would be incurred with thin slabs because casting speed can be increased, thus allowing also an improvement in the surface of the product. The capital cost of a plant for the casting of thin slabs is likely to be less than that for the casting of thick slabs, mainly because the machine need not be so tall. It is believed that fairly high tonnage rates, say 40 or 50 tons/h/mould, could be achieved on these plants.

Most of the continuous-casting work undertaken to date has been done with killed steels, although rimming steels have been cast in notable cases, both in the USSR and in the Inland trials at Atlas Steel, Canada. We are not at all certain whether rimming steel in many of its applications will be required if continuous-cast killed metal is available, but where the need for rimming steel as such is essential we have no doubt that it can be produced by continuous-casting methods. We expect, however, that it will be preferable to cast rimming steel in billet sizes greater than 6-in. square and in slab sections thicker than 3½ in.

Before leaving the question of economy I think I should make it clear that one cannot really judge the whole picture until experience is gained with a works designed for production around the continuous-casting process itself. Continuous casting introduces a number of new conditions. Ideally, it should be matched with continuous steelmaking. The fact that the product is delivered continuously, cast to sizes already approaching the finished section, and sufficiently hot and of sufficiently consistent quality to allow it to be processed immediately to the finished requirements, gives scope for savings in reheating and production time. If production plants based on the full-scale use of continuous casting can be designed to take advantage of these savings, the margin of economic advantage is likely to be very considerable.

The Chair was then taken by Mr **A. H. Ingen-Housz** (Past-President).

There followed the showing of film on continuous casting in the USSR, at William Jessop and Sons Ltd, and at Barrow Steel Works Ltd.

General discussion on the following papers then took place (dates of publication in the *Journal* in parentheses):

"Continuous casting at Barrow" by *I. M. D. Halliday* (1959, vol. 191, pp. 121–163)

"Continuous casting of steel in the USSR" by *M. S. Boichenko, V. S. Rutes, D. P. Evteev,* and *B. N. Katomin* (1959, vol. 191, pp. 109–121)

The papers were presented respectively by Mr Halliday and Mr Boichenko. Mr Boichenko's presentation and replies to questions were interpreted by Dr S. Klemantaski.

Dr **B. H. C. Waters** (K & L Steelfounders and Engineers Ltd): I should like first warmly to congratulate Mr Boichenko and to comment on his paper. He describes much excellent work done in the USSR using continuous casting on a large tonnage, production scale. Achieving large tonnage production by this process presents problems and his comments on how these problems have been overcome in the USSR would be interesting. It would be interesting to hear what Mr Boichenko has to say about the relative advantages and disadvantages of lip and bottom pouring in continuous casting. Mr Halliday uses lip pouring, but bottom pouring seems to be more generally favoured in the USSR. Mr Boichenko's comments on mould design, particularly in relation to the use of reciprocating moulds and moulds tapered inwards at the bottom, would also be interesting. I would also like to ask what progress has been achieved in the USSR with rimming steels. Finally, would Mr Boichenko elaborate, on the basis of tonnage operation, his remark that the quality of continuously cast material as compared with that produced

by standard ingot casting has been judged in favour of continuous casting.

Turning to Mr Halliday's excellent paper, I would first congratulate him and his colleagues on what appears to be a classic work. Also I would presume to congratulate the directors of the United Steel Companies for their steady faith in what must have been a rather expensive and sometimes, apparently, a rather speculative piece of work. It has been crowned with much success, to which the combination of small cast section size and high casting speeds has contributed considerably.

I would first ask Mr Halliday for comments on the negative strip mechanism and particularly whether he considers it essential to the process. Also, does he consider the use of dip couples in the ladle to be entirely reliable, or may convection currents occur ? Are any special steelmaking techniques used, or is nothing more done than the evident close control of the tapping temperature ? Does Mr Halliday consider that closed pouring might be useful in a system such as his ? Also how did Mr Halliday overcome the nozzle blockage by alumina which he mentions ?

Finally, commenting on both papers, both authors are optimistic when considering the future of continuous casting in relation to strip mills. Do they consider that the time when strip mills will be markedly affected by the use of continuous casting is near or far away ? Also in a way the rather different lines of development depicted in the two papers appear to be leading to slab casting at speeds of about 7 ft/min. Do the authors consider that this is going to be the most commonly used continuous casting plant in the future ?

Mr **Halliday** (in reply): The first question to me was: Is it essential to have negative strip with continuous casting at high speeds ? The answer is simply ' Yes.'

Next, is the thermocouple method used for checking the temperature of the metal in the ladle accurate and reliable ? Our thermocouples are reliable when, if maintained properly, they produce a satisfactory temperature-rise chart. Thermal currents do not seem to have very much effect. They may exist but, if so, they exist due to very small differences in temperature and introduce no complicating effect.

If temperatures are checked accurately at the furnace, the ladles arrive at the machine with metal only a few degrees above the required temperature, and some minutes later the metal is within 2–3° of the casting temperature. I do not think we can get much better checking. The thermocouple is always dipped in roughly the same position, which also helps.

Can steel be made for our type of operation by virtually the orthodox methods ? The answer is ' Yes.'

How do we view closed pouring, and has it been tried ? This has been tried but we are probably best advised to omit this question from this meeting, for it opens up quite a different field. Although the historical survey given as an Appendix to my paper covers many things, it is not complete by any means. We have tried closed pouring, casting both vertically and horizontally. At this stage I would say that it is best to forget closed pouring for steel.

Blockage of the nozzle by alumina occurs in certain cases, for example, when casting silico-manganese steel. This trouble has nothing to do with the design of the non-swirl nozzle and we have greatly alleviated the trouble by using a larger nozzle. We have also alleviated the proneness to blockage very considerably by using ferrosilicon with a specified aluminium content of less than 0·5%. However, if we were to double the length of our casting period we should probably again run into a tendency to blockage. The use of a larger nozzle allows running for a longer period before blockage becomes apparent. Generally, we are able to cast the whole 7½ tons of silico-manganese steel at high speeds, well above 200 in./min, without this interference. This reflects the relatively good control of flow available, allowing operation with a much larger nozzle than we should normally need.

Mr **Boichenko** (in reply): I have managed to write down eight questions.

First, our views on the relative advantages of lip and bottom pouring ladles. Our experience, especially with large sections, is that bottom pouring is far better. As to smaller sections, Mr Halliday has more experience than we have, and for the smaller sections and alloy steels perhaps lip ladles are essential. On the other hand, work in Canada suggests that stopper ladles should be used, and this confirms our experience.

With regard to the relative advantages of stationary and moving moulds, we feel that the final answer to this question will be available only when we have accumulated some experience with large industrial installations. At present both types of installation are working with equal success. At the Krasnoye Sormovo works a fixed mould is used. Naturally, we all understand that when using fixed moulds special attention must be paid to lubrication.

On the question of using moulds tapering towards the bottom, our experience suggests that this is of great practical importance. This effect may be due to the reduction or elimination of the influence of the air gap, and some increase in the speed of withdrawal of the slab or bar may result. We intend to use this mould construction for the larger slabs which have been mentioned in our report. I must emphasize that all I have said refers only to built-up moulds.

I turn next to the question of the quality of continuously cast steel and its advantages. There is hardly time to speak at any length on this very important question. We can, however, say that with none of the steels which we have worked has there been any appreciable deterioration due to continuous casting. With some steels, especially transformer steels, some improvement in cold plasticity has been noted, and we are at present studying this question in some detail, especially in connection with the installation being built at Novo-Lipetsk.

The next question is that of rimming steel. This, again, is a very large and complicated question, and we have not enough experience so far to be able to give an unambiguous answer. The experience that we have suggests that it is more difficult to cast square sections, for example, a 200-mm square, on the surface of which a considerable quantity of subcutaneous porosity appears, and we are studying this question at the moment. With rimming steel slabs, however, our experiments have been more successful. It is possible that this is connected with the dimensions of the sections and the associated flow of liquid metal inside the slabs.

Next, the speed of casting of slabs. There is no doubt that the rate of casting of slabs is considerably less than for small sections. It would indeed be wonderful if we could attain with slabs the speeds now being attained by Mr Halliday with his small sections. We are sure that with the sections mentioned in the paper the rate will be 0·5, 0·6, or 0·7 m/min for carbon steels. For 18–8 stainless steel containing titanium, 150 mm × 500–600 mm slabs, we obtain speeds of 1·2 or 1·3 m/min. Our preliminary calculations suggest that such speeds are quite sufficient even for 260-ton ladles.

Finally, on the prospects for continuous casting in connection with continuous strip mills I think that all workers in the field believe that the introduction of continuously cast slabs for this type of rolling is the first

development, and we think it can be used for both hot-rolled plate and cold-rolled thin strip of carbon steel and low-alloy steel. We think that even the preliminary economic calculations are sufficiently favourable for this problem to be solved.

Mr **T. G. Marple** (Barrow Steel Works Ltd): I should like to make a few practical observations on the Barrow work.

We are preparing to put the twin-strand machine into full production. It will start next year on three-shift working making ordinary commercial products. In the meantime we shall mechanize the delivery side, introducing mechanized cut-off torches and installing mechanically operated hot banks for collection.

Mr Halliday stressed the importance of negative strip for high speed continuous casting. Negative strip prevents fracture of the cast billet skin and, in practice, it is essential that maintenance should be held at a high standard. The actual difference in movement is only about $\frac{1}{32}$ in. on a 2-in. stroke, and this amount can be lost if wear in the moving parts is allowed to take place.

The plant at Barrow is a dual purpose one. We are constructing a large slab machine on the side you could not see in the film. Arrangements have been made for the ladle, in its cradle, to be tilted in both directions, so that metal may be teemed from it either towards the twin-stranding side or towards the slab side.

A roller apron and sprays are supported immediately below the mould, which is designed to cast 36 in. × 5½ in. slab sections.

The withdrawal roll unit consists of four rolls, arranged in two pairs. Spring loading can be applied on the necks of the adjustable rolls through motorized roll drives, whilst the roll stand has been designed to accommodate slabs up to 36 in. wide and also two 9-in. square blooms. Below the rolls a cutting torch can be pushed into position. As we are limited to 7¼-ton charges, there is no necessity to cut the slab during casting. This slab machine will be in production very shortly. It is due to start in the New Year.

Mr **G. Wilz** (Etablissements Cail): I should like to ask Mr Boichenko a question about rimming steel. Is there any difficulty in the continuous casting of rimming steel in maintaining the level of the steel in the mould?

Mr **A. J. K. Honeyman** (Steel Company of Wales): In casting wide slabs for strip mills, shape is of paramount importance, but not much has been said on this subject. Would Mr Boichenko state whether this problem of shape has been solved.

Mr Boichenko also referred to the waviness which he has experienced due to distortion of the mould. Could this be caused by the rather long mould used? He states that the higher the ratio of slab width to slab thickness, the greater is the danger of longitudinal cracks along the wide face. What in his opinion, is the optimum width/thickness ratio?

Table II shows the mechanical properties of continuously cast material of different compositions, but in every case the tensile strength is about the same. It is certainly very surprising to get a tensile strength of 38 tons/in² in normalized 0·22%C steel. Would Mr Boichenko please state the diameter of the test piece and the gauge length used to determine the elongation.

Dr **E. Brewin** (BISF): First, could Mr Boichenko give us any idea of the future plans for expanding continuous-casting capacity in the USSR? We have had various reports in the UK of an enormous plant now being considered and discussed. If that is so, perhaps some details of the tonnage and siting are now available.

Second, there have been various reports in the UK of production targets by this process by 1962–63. One figure in general circulation is 2 million tons total production by 1962–63. Another figure has been given, that of 10 million tons proposed production capacity for 1968. Perhaps Mr Boichenko could tell us whether this is correct.

Mr **G. T. Harris** (William Jessop and Sons Ltd): First, is there any Soviet experience on the continuous casting of high-alloy steels, particularly high-speed steels and tool steels?

Second, has Mr Boichenko any experience on whether the flat surfaces of squares and rectangles remain true, or whether there is distortion of such sections?

Mr **Boichenko** (in reply): On the question of maintaining a constant level of metal with rimming steels, we have not encountered any difficulties.

On the question of the optimum thickness/width ratio of slabs, we use a flat shape even without round corners. This is of great importance with our built-up moulds with regard to the construction of those moulds. For carbon or low-alloy steels this does not introduce any difficulties in rolling either for squares or for slabs, although rolling-mill operators were at first very hostile to this. However, for some alloy and high-alloy steels some curvature should be produced at the corners to ensure satisfactory further work. Although we have no direct experience of rolling slabs in planetary mills, we think that for this type of mill there should be sufficient roundness towards the thin sides of the slab.

As for the optimum thickness/width ratio, we think that 1 : 5 would be quite satisfactory under present conditions, but we think that it would be possible to obtain further reduction. This is especially important for wide hot- and cold-rolled strip, in the direction in which rolling-mill practice is evolving nowadays strip widths of over 2 m being required. We have as yet no final answer to this, but we think that with the aid of roller cooling after the mould such a ratio of side lengths will be possible. We think that the main difficulty in obtaining large dimensions with small thicknesses is due to the actual difficulty of pouring the metal into such sections.

With regard to the results of work with rimming steels, I said that the casting of slabs was easier than the casting of squares. It is at present difficult to explain this, but it shows in practice. We hope to solve this during 1959.

As to the productivity of future installations and the total production in 1962 and 1968, the development of installations is dealt with in our paper. We think that the Novo-Lipetsk installation with its 80-ton ladle and the Stalino installation with its 140-ton ladle will in the next few years give sufficient information to decide on the further application of this process, especially in sheet rolling. The possibility of an installation for a 270-ton ladle is not excluded for the production of hot-rolled strip from slabs whose dimensions have already been indicated.

On the question of high-alloy steels and the distortion of shape in continuous casting, in our paper all the steels with which we have worked are described. Most of our experiments were carried out with 18–8 titanium-stabilized stainless steel. We have not worked with high-speed steels.

We have hardly observed distortion with slabs, but for example with 200-mm squares we notice some effect. We have not so far found out the main cause of this, and we should be very grateful to receive any information which members of this conference have on the subject. In this connection we should like especially to hear Mr Halliday's comments because of the successful operation of his installation.

Mr **Halliday** also wrote in reply: In regard to the future effect of continuous casting on strip mills, I agree entirely with Mr Boichenko's reply to this question and can only

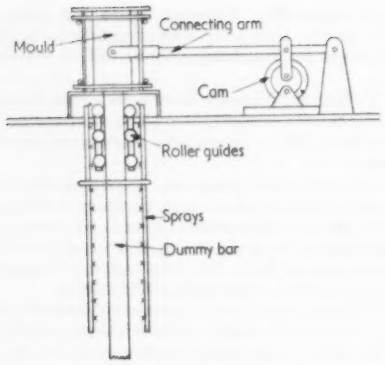


Fig. A—Method of reciprocating the mould

add that his views have already been submitted to the meeting both in his own paper, under the section headed 'Economy,' and in Mr Wingate's opening review. However, the likelihood of a large continuous-casting installation being made to supply a sheet or strip mill in the near future seems to depend mainly on the need arising in the first instance (at least in the UK), and, in addition, the existence of a sufficiently strong spirit of enterprise to take up such a project for the first time. There seems little doubt that the remaining practical problems can be solved and the economics are certainly promising.

The subsidiary question asked indirectly by Dr Waters in his suggestion that slab casting would eventually be at speeds of about 7 ft/min was difficult to answer. There appears to be no particular importance or virtue in the figure 7 ft/min. Casting speed for slabs will be related in future, within broad limits, directly to the height of the machine (particularly spray zone height) and, inversely, to slab thickness. For any particular installation site conditions and requirements, both on the metal supply and product delivery sides, have to be taken into account and each project needs to be considered as an individual case.

In regard to distortion of cast shape, experience at Barrow has shown for small square sections that a fairly regular square section could be repeatedly obtained, even at high speeds. In certain trials the square section was found to be rhombic in shape and in all these cases the cause of this was traced to the mould. Although the whole picture cannot be filled in, it seems that certain moulds can become distorted in operation. The difficulty, whilst important, does not hinder production and in the case of the long-life moulds for 2-in. square billets referred to in my paper has apparently been completely eliminated. Work is, however, proceeding in the case of other mould designs.

For small slab sections cast at high speed, another cause of distortion arises. There appears to be a relationship between casting speed, spray intensity and slab section shape. If the former two variables are incorrectly controlled the resulting slab shape can easily be poor, although internal structure is still reasonably satisfactory. Correction of casting speed for a given spray intensity or, *vice versa*, can immediately, during a run, restore good shape.

Part 2 Afternoon session

At the afternoon session Dr **J. Pearson** (BISRA) showed a film made at the BISRA Laboratories in Sheffield by the research staff of the BISRA experimental plant for continuous casting.

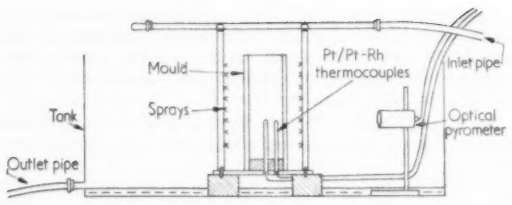


Fig. B—Apparatus used for studying secondary cooling

Mr **G. Fenton** (BISRA), asked by Dr Pearson to describe how the experimental plant had been used to answer commercial problems, said: This brief report describes a typical example of the use of an experimental plant to determine the basic requirements for a production machine. It is also to some extent in the nature of a postscript to the film which has just been shown, and to the paper which we presented at Charleroi, bringing them both more up to date.

About a year ago, BISRA was asked to carry out a full investigation into the continuous casting of silicomanganese spring steel to En45 specification. Certain requirements had to be met. First, the steel had to be made in 3-in. square section. Second, because of considerations relating to the number and size of the furnaces, it was necessary to cast at a speed of 12 ft/min. The third requirement was that the quality of the product had to be satisfactory with regard both to internal structure and to surface condition. We had to determine the conditions under which all these requirements could be met simultaneously.

The plant as it existed at that time has been shown in the film. It was not fitted with a secondary cooling zone, and it had in fact only just become possible to apply secondary cooling after some extensive modifications to the water system. It is, I think, obvious that we could not even contemplate casting at 12 ft/min without efficient secondary cooling, so that one major consideration was to determine the optimum conditions for this zone.

The second major consideration concerned the mould system itself. Up to that time we had worked with spring-mounted moulds of the general design shown in the film, and we found that such systems worked very well indeed, contrary to some of the opinions expressed at Charleroi. The theory of spring mounting is concerned with the prevention of skin rupture in the mould, and we are convinced that once the correct spring characteristics are established there can be no doubt about the validity of that theory. The proof lies in the ingot surfaces produced regularly from these moulds at the BISRA Laboratories. We had to remember, however, that these spring mountings had been developed for use with relatively low casting speeds, of up to about 6 ft/min for 3-in. square section. This raised what I have called the second major consideration: would the system work satisfactorily at a much higher speed?

To deal with these two immediate problems we started two series of experiments simultaneously, one on the casting machine to investigate the mould system and the second on the melting shop floor to determine the optimum secondary cooling conditions.

For the series of casting experiments on the machine, we first installed a temporary spray zone. At that time we were not primarily concerned with the effect of that zone on the structure, but simply wanted to ensure that the ingot could leave the mould at 12 ft/min without break-out, so that we could study the mould action and the resulting surface quality. This temporary zone was

quite successful in that respect. We started casting with the spring-mounted mould, and it soon became obvious that the spring mounting, which was satisfactory at the lower speeds, was not sufficiently sensitive to prevent skin tears at high speed. This meant frequent stops of a few seconds duration to allow the tears to heal. That was not surprising, because if we extrapolate the theory of spring mounting to these high speeds it is clear that we shall need a softer, well damped system, allowing greater mould movement.

We knew, however, that to develop such a system would require considerable experimentation, and, since this work was urgent, we decided to install a second type of mould suspension, one which we knew as 'compression–release'. We applied for patent protection for this idea early in 1954, and shortly afterwards found that Mr Halliday and his colleagues at Barrow had been independently thinking along the same lines and were using a similar system on their own plant. The system which we know as 'compression–release' is almost identical in principle with the 'negative strip' system used at Barrow with such outstanding success. With our device the mould is reciprocated over a distance of 2 in. and the cycle is such that the mould moves downward at a speed of 13 ft/min and then returns upward at four times that speed. Figure *A* shows the method of conversion of the mould assembly. The mould tube (which remained 19 in. long) and its water-jacket were built into an accurate guiding frame and provided with trunnions which engaged with a connecting arm, pivoted at its far end. Reciprocation was made by means of a cam which was so positioned and contoured as to give the desired mould cycle.

In practice this system has worked very satisfactorily. The first and most obvious thing that we noticed was that we were immediately able to cast without any of the interruptions which had proved necessary with the spring mounting. High-speed cine-photographs taken during casting confirmed that skin rupture in the mould was very infrequent indeed. Initially, however, we were not happy about the ingot surfaces; at regular intervals along the length they showed marks very similar to those already described by Mr Halliday. We found that the severity of these marks was largely a function of the smoothness of the mould cycle. The same high-speed ciné photographs showed that in the early casts the cycle was not so smooth as we wanted it to be; but by careful attention to the engineering of the system to remove backlash, and particularly by installing dampers which smoothed out the top and bottom of the stroke, these marks became much less severe. We still find occasional patches of more severe marking, which we are sure are due to inefficient oil distribution. We are now engaged in improving the oil distribution and at the same time in exploring a range of mould cycles and amplitudes.

Turning to the second series of experiments on secondary cooling, we set up a simulation technique on the shop floor to avoid time-consuming and very costly trials on the plant. Figure *B* shows the apparatus used. The steel was melted in a small high-frequency furnace and cast into a small 3-in. square collapsible mould, itself mounted on a refractory base. From this base, two thermocouples projected into the mould cavity, one axially and the other midway between axis and wall. A spray assembly was located outside the mould as shown. The procedure was to cast the ingot from the high-frequency furnace, top up the mould with vermiculite, and then remove the mould walls at a time corresponding to that at which the ingot would emerge from a continuous casting mould. Immediately after removal of the mould, the sprays were turned on for a predetermined time, and reciprocated gently to simulate

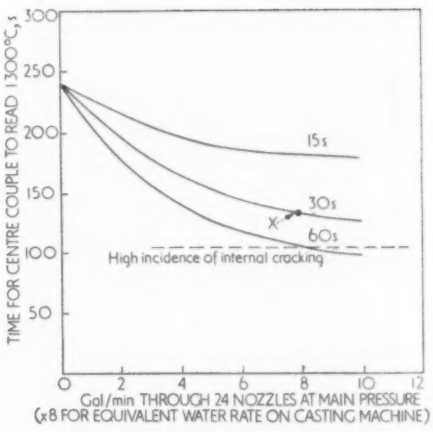


Fig. C—Variation of freezing time with rate and duration of spray cooling

movement of the ingot through the spray zone. When the sprays were turned off, a radiation pyrometer recorded the surface temperature.

In this way we obtained a series of cooling curves which gave us information which could be used to decide the position and speed of the cutting torch. The most important part of this information is summarized in Fig. *C* which shows the time required for the centre of the ingot to reach 1 300° C and the variation of this time with the intensity and duration of the water spraying. We chose the temperature of 1 300° C because it seems to be about the true solidus temperature for this particular steel, although admittedly this is very difficult to determine accurately. We feel that a higher figure, say 1 350° C, might be taken but 1 300° C is on the safe side.

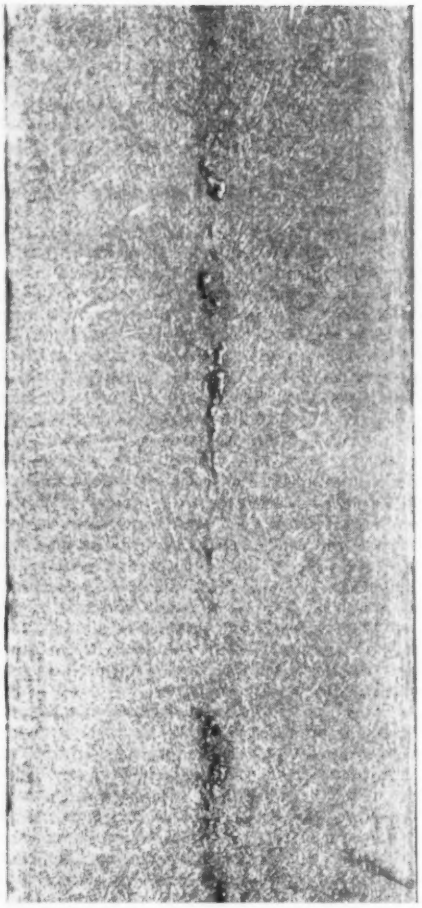
All the ingots cast in this way were sectioned, so that we could study the metallurgical effect of the water spraying. In particular, we found that any attempt to reduce the freezing time below 100 s resulted in internal stress cracks. To allow a safety margin, we decided to adopt conditions giving a freezing time of 135 s, and for convenience took the point *X* on the centre curve. This corresponded, on the plant, to a spray zone 6 ft long and with a flow rate of about 60 gal/min.

At this point we dovetailed the two series of experiments together on the plant itself and built in a spray zone of this type. From this point we could look into the important questions of internal quality and permissible casting temperature range. Because of our limited furnace capacity at BISRA our casting times are quite short, and we find that there is a negligible temperature drop during casting. By varying the tapping temperature, we can easily obtain a range of casting temperatures on the plant. The temperatures quoted below are metal temperatures in the tundish in each case.

We found that the lower limit of casting temperature was 1 460° C; below that, the tundish nozzle would freeze. That was not unexpected as 1 460° C is roughly the liquidus temperature of this steel. One would expect the upper limit to be set either by the tendency to breakout or by unsatisfactory central structure. In practice we found that it was a tendency to breakout which limited the upper temperature, and we put this at about 1 510° C, giving a permissible temperature range of the order of 50° C.

As might be expected, the degree of superheat has

(a)



(b)

Fig. D—Longitudinal sections of ingots cast at (*a*) 1 470° C, and (*b*) 1 505° C

an effect on the ingot structure. The two sections shown in Fig. *D* are cast roughly at the upper (on the right) and lower (on the left) temperature limits; with the higher temperature the ingot shows somewhat more central unsoundness and a coarser dendritic structure, but in neither case is there any significant segregation. Although the higher-temperature ingot has more central porosity, it rolled out satisfactorily. The porosity virtually disappeared at 50% reduction in area and completely disappeared at 70% reduction. We did not find any effect of superheat on surface quality.

That, in brief outline, is an example of the use of an experimental plant. Undoubtedly there are points which can be fully investigated only on a larger unit, casting greater quantities of steel, but nevertheless a small experimental plant can, as we have shown, provide all the basic data required for the design of a production unit.

The following papers were then presented (the dates of publication in the *Journal* are given in parentheses):

"Solidification characteristics of continuously cast low-carbon steels" by *M. Tenenbaum, C. F. Schrader,* and *L. Mair* (1959, vol. 191, Jan., pp. 20–33) presented in the absence of the authors by Dr J. Pearson (BISRA).

"Continuous casting of high-speed steel" by *B. H. C. Waters, W. H. Pritchard, A. Braybrook,* and *G. T. Harris* (1958, vol. 190, Nov., pp. 233–248) presented by Dr B. H. C. Waters. A general discussion on continuous casting followed.

Mr **A. J. K. Honeyman** (Steel Company of Wales Ltd): Mr Boichenko has kindly presented me with a copy of his book on continuous casting, but unfortunately, it is in Russian. It is time for a really good textbook in English on this subject. The papers presented today, together with the references therein and the discussion arising out of them, should provide excellent material for such a book.

I have two general points to raise. First, I do not understand why mould length should vary from 19 in. in the case of the BISRA mould up to 50 in. for the Russian mould. Will the authors please give their reasons for the mould length which they employ.

Messrs Speith and Bungeroth use the expression "If basic Bessemer rimming steel also can be successfully cast by the continuous process . . ." What success have they achieved and what in their opinion are the prospects for complete success ?

Messrs Krainer and Tarmann referred to metallurgical difficulties due to hydrogen. Will they please explain what exactly these difficulties are.

Turning to the paper by Tenenbaum, Schrader, and Mair, it is interesting to note that the mould addition of aluminium is apparently very critical, a difference of

0·2 lb/ton resulting in a change from no solid rim to one 2·5 in. thick. The authors suggest that "the accumulated head of liquid metal is sufficient to suppress the carbon–oxygen reaction to the extent that gases no longer freely evolved." May it not be that the prime cause of suppression of the carbon–oxygen reaction is the thinness of the liquid core preventing convection currents, as indicated in the paper by Krainer and Tarmann? Reference is made to skin blowholes, and an explanation is given that this is due to very rapid initial freezing. Could this be overcome by devising a means of slower cooling before the metal came in contact with the water-cooled copper mould? Alternatively, would it be possible to circulate warm water, so as to reduce the chilling effect? Would a longer mould also help in obtaining a thicker rim, since the rate of cooling would be slower than that in the spray cooling zone? Could the authors state if there are any other causes of surface defects? Finally, would the authors state from their wide experience, what in their view are the prospects of overcoming the numerous problems enumerated by them.

Dr **C. Herrmann** (Demag, AG): A few months ago the metallurgical plant at Terni in Italy put into operation a Demag continuous steel-casting plant built to the Junghans system. Messrs Mannesmann at Huckingen and Böhler at Kapfenberg are taking care of the metallurgical problems. The industrial plant manufactured for Terni is designed for casting eight strands with a cross-section of $4\frac{1}{2}$ in. × $4\frac{1}{2}$ in. and $5\frac{1}{2}$ in. × $5\frac{1}{2}$ in. from 60–70-ton melts from an open-hearth furnace, and for casting four strands from the 35-ton electric arc furnace melts. The yearly output is rated at 180 000–200 000 tons. The casting speed is up to 8 ft/min, and the cut-off bars have a length of 10 ft.

The plant consists of a framework with platforms on five levels, into which the eight casting stands are built in two groups of four units each. For each group a control cabin has been built on the fifth platform, which is 43 ft above the floor. The bottom-poured ladle with the molten steel is emptied into a tundish from which the steel flows into the eight moulds. The tundish can be moved and tilted hydraulically. The water-cooled copper moulds consist of sections which are bolted together and can be separated, cleaned, and dressed whenever necessary.

An important factor in obtaining a satisfactory strand surface is the smooth gliding of the skin of the metal over the mould wall. The injection of sufficient oil immediately at the mould wall prevents oxidation, and a constant upward-and-downward rocking of the moulds is a further means of reducing the danger of the metal adhering to the mould walls. The oscillations can be adjusted from 0 to 1 in., and the rate of rocking can be infinitely varied between 30 and 100 times/min.

The cooling chambers below the moulds have double walls and are also fed with cooling water. Each chamber has four recesses containing a distributor with water-spraying nozzles. While the strands run through the cooling chambers they are guided, supported, and kept in shape by rollers. During the subsequent drop, which extends over several metres, the strand is allowed to travel without guidance, so that it is exposed to the air for further cooling and a perfect internal structure is obtained. Two feeder rolls then grip each strand. The rolls are mounted on arms which can be opened out and closed by an hydraulic cylinder, and the strands are withdrawn from the bottom of the moulds at speeds which can be varied between 0 and 8 ft/min.

The strand, which has by now solidified through to the centre, is then cut up into bars. The flame-cutting machine is mounted on a carriage running on a vertical column at the speed with which the bar is withdrawn. On the same carriage hydraulically operated tongs are fitted which grip the bar as it is cut off and lower it into quiver-like receptacles. These are tilted into the horizontal position with the cut-off bars, which are discharged on to and removed by the roller conveyor.

A second and similar plant for the casting of four strands, 180 mm × 180 mm and 220 mm × 220 mm, is now being built for a French integrated iron and steelworks.

The total capital investment for the Terni production plant, including foundations and erection, amounted to about £500 000, which is only a fraction of the cost of a blooming mill.

Mr **J. Zaeytydt** (CNRM): The CNRM plant at Charleroi at the moment is equipped for casting two strands 100 mm and 200 mm square with provision for two more strands, full details of which have been given in my paper.

Several papers have been written dealing with the continuous casting of low-carbon steel, but in all these cases the steel was produced in electric or open-hearth furnaces. At CNRM, our main interest is to cast low-carbon basic Bessemer steel, killed and rimming, in 20-ton charges.

We find from our experience that although this work on low carbon steel made by the electric or O H process is most interesting, when one attempts to apply it to the basic Bessemer, one finds great difficulty in the temperature control.

In basic Bessemer process there is a considerable drop of temperature from the end of blow to the actual pouring temperature. A drop of 40–80° C is not unusual in a 20-ton converter for end temperatures of 1 590–1 660° C. As one can see, in the electric process when a tapping temperature of 1 620° C is quoted, using basic Bessemer, one would need an end of blow temperature of between 1 660–1 700°. (A normal basic Bessemer end point temperature could be 1 600–1 630° C.) These high end point temperatures are not a commercial proposition, they may affect the nitrogen, oxygen, and phosphorus contents also the iron content of the slag will increase.

To date we have successfully cast 2-ton charges with a teapot ladle preheated to 1 500° C and steel temperature at the end point of 1 650° C. The temperature of the steel in the tundish bath was 1 560° C which we consider is a minimum necessary for passing through the tundish the full charge without premature freezing. (The liquidus–solidus zone starts at 1 520° C.)

When we attempted to take a charge of 10 tons with a 20-ton bottom-poured ladle, with normal preheating, steel temperature 1 650° C at the end of the blow, the resulting steel temperature in the tundish was 1 520° C. Under these conditions of no superheat in the steel the tundish nozzle makes up within a few minutes, although the tundish nozzle was at 1 530° C and the bath at 1 475° C (preheat temperature).

Efforts are now being made to reduce the heat loss by higher preheat temperature in the ladle, by replacing the addition of solid ferro-manganese by liquid ferro-manganese and by reducing the heat loss at the converter. With a stoppered ladle we consider 1 200° C as the maximum preheat possible with efficient functioning of the stopper. This limitation, however, does not apply to a teapot ladle which we are also considering trying.

Mr **J. Savage** (Continuous Casting Co Ltd): Of the papers presented this morning, that by Boichenko and his colleagues is in my opinion one of the most valuable we have seen on this subject, covering as it does the basic theory, engineering, existing practice and prospects for continuous casting in the USSR. I should like to make a few remarks on the final sections of this paper, which

Fig. E—½-in. dia. bar extruded from continuously cast high-speed steel

deal with heat exchange and with technical and economic factors.

Some four years ago a number of papers on the subject of heat exchange and the principles of mould design were published in this country. I was particularly interested to see that the basic conceptions and the quantitative results which were then put forward as a result of the work carried out by BISRA have been confirmed in this Russian paper. I feel sure that the statement which will create the greatest impression is that careful consideration is now being given to replacing blooming and slabbing mills by continuous-casting machines in new plants with outputs of up to five million tons per annum. It would be interesting to know what the Russian authors would do if they had the responsibility now for bringing a new hot strip mill into commission in five years time, not for five million tons, but for an initial output of one million tons, capable of expansion in the normal way.

Turning to the papers more generally, one finds on analysis that the results reported show that the relation between casting speed and size of section, suggested by Speith and Bungeroth, fits current practice very closely for a wide range of sections. The relation can be extended to show that the casting rate in tons per hour for one casting strand is of the order of three- to four-fifths of the perimeter of the section measured in inches. For example, taking extreme cases, a 2-in. square ingot is apparently cast at about 6–7 tons/h, while 24-in. × 6-in. slab is cast at about 40 tons/h. This relation fits into a simple pattern the widely-scattered data on casting rates given in these papers.

What is not clear is the nature of the problems which had to be solved in casting billet sizes. It is worth examining these, so that the degree of success achieved and perhaps some practical limitations in the casting of small sections can now be more readily appreciated. Barrow, Jessops, and BISRA have all been confronted with the problems of keeping a small tonnage of steel in the ladle liquid for long enough to cast at rates which are very small by conventional standards. This challenge has been met, and the excellent films which we have seen show the solutions which have been found.

Looking at the matter more broadly, perhaps it is possible to summarize the position in the UK by suggesting that almost all the efforts have been concentrated on the casting of small sections, from 2 in. to 4 in. square. There was certainly a great need to succeed in this field, as the range of application was potentially very wide, but we must recognize the nature of the problem which had to be faced. The need to solve these problems has been responsible for the apparently slow progress of continuous casting in the UK. The fundamental and applied work which has been done has, of course, resulted in the acquisition of very considerable experience and know-how.

We must get one fact very firmly established in our minds. The casting of so-called billet sections on the one hand and of blooms and slabs on the other are almost completely separate fields. There can now be little doubt that the latter is far easier when judged from the very important aspect of liquid steel pouring rates. Perhaps this is the reason which has led people in France, Canada, Austria, Germany, and Russia to concentrate their main efforts on the larger sections, with apparently more spectacular results. I consider, however, that such a view would be entirely superficial and that we should not underrate the very considerable progress which has been made in this country in the most difficult field of casting small billet sections.

Mr **G. Hoyle** (BISRA): I am a little puzzled by the nature of the centre segregates in continuously cast high-speed steel. In the paper by Waters *et al.* reference is made to carbide segregation, but I do not think that this is entirely responsible for the phenomena shown in the illustrations. In Fig. 28 of the paper it is apparent that the actual carbide colonies represent only a small part of the white-etching central segregates, although there are rings of high carbide concentration within these areas. The authors also suggest that the banded segregates in bar from continuously cast high-speed steel are due to carbide segregation.

We have examined and extruded to ½-in. dia. bar a small quantity of this material, and on this one sample it is difficult to tie-up carbide distribution with the central band. Figure *E* shows this. All the photographs are of the same specimen. The top section (*a*) is from an enlargement of a low-power macrophotograph, and the central light-etching band is very well defined. The centre strip (*b*) shows the same specimen photographed at higher power with macro illumination; we still have the banding, but it is less pronounced. For these two exposures the specimen was etched in nital. The bottom strip (*c*) shows the result of a different etch which reveals carbide structure but not the banding. This strip was made up of 13 exposures on 35-mm film at a magnification of 30. The extent of carbide segregation, as shown by the stringers, does not seem to be excessively severe, and it does not seem that we can find any relationship between the banding and the carbide stringers. Personally, I think it more likely that the effect is due to porosity, as I have seen somewhat similar results on other high-speed steel bars extruded from unsound ingots.

We have examined the hardening characteristics of the steel and find that, as shown in Table I, the outer zone of the bar is consistently harder, by about 30 VHN, than the inner zone after full secondary hardening. On such specimens there is a ring of segregate at mid-radius position, separating the two zones. Again at first sight these look like carbide segregates, but high-power examination suggests that they are areas of retained austenite associated with smaller carbide segregates. This austenite is unusually difficult to break down by multiple tempering, suggesting local alloy enrichment; but, like the authors, we have not observed any significant segregation either by spectrographic or by X-ray fluorescence techniques. Is it therefore possible that some gaseous element is the cause?

We have examined the properties of the inner and outer bands by the BISRA band test, and the results are shown in Table 2. Again the inner zone is softer than the outer, but the limit or proportionality is the same in both cases (each figure quoted is the average of six tests) and the apparent breaking stress is higher at

TABLE 1—V H N of bar extruded from continuously cast high-speed steel

Condition	Outer band	Inner band
O Q 1 280° C	765	773
plus T 560° C 1 h	880	846
plus T 550° C 1 h	873	826
plus T 565° C 1 h	876	830

T = tempering

the centre. The remaining figures are empirical values developed by BISRA* for the express purpose of comparing tool-steel properties. Without discussing these in detail, we can say that the inner portion has superior ductility and toughness, of which we believe the quantity Q to be a reasonable assessment, and superior tensile strength or true breaking stress. This seems rather an odd result, in view of the allegations made against continuously cast high-speed steel. Magnetic tests have failed to show any appreciable difference between these samples. The inference is that, in spite of the undesirable effects in the macrostructure, the centre of continuously cast high-speed steel is apparently no worse, and may in fact be better, than the outside.

Mr **M. D. J. Brisby** (W. S. Atkins and Partners): The discussion has dealt very fully with the metallurgical and engineering aspects of continuous casting but little has been said about capital and operating costs. I think the time is ripe for plant builders to undertake what may prove to be a most worthwhile exercise, project studies of continuous-casting plants of various sizes, comparing capital and operating costs of continuous-casting installations on the one hand and conventional plants on the other. May I suggest that this should be done for plants with annual capacities of 100 000, 200 000, 500 000, and possibly even 1 000 000 tons.

Rolling mills do not show a good return on their capital costs until they are run at something approaching full capacity. In the case of a billet mill, this might be in the region of 750 000 tons a year. Continuous-casting plants may be economic at much lower levels of output. At all events it should be possible to find a cost break-even point between continuous casting and conventional processes and so determine the range in which continuous casting shows up to best advantage.

Mr **B. Tarmann** (Gebr. Böhler): Referring to the paper by Dr Krainer and myself, Mr Honeyman asked what were the metallurgical difficulties due to hydrogen. One of them concerns the mould. Hydrogen is formed by interaction between the hot steel and the water which is likely to condense along the wall of the mould; it gets dissolved as H_2-gas and causes porosity if escaping during solidification. To avoid condensation of water in the mould, we have introduced a cooling system which allows us to maintain the temperature of the circulating water a few degrees above dew point.

Mr **J. S. Morton** (Continuous Casting Co. Ltd): The question has been raised of the capital cost of continuous casting installations. This is a very big subject, and, while these studies must be made, they are extremely difficult and complicated, because the capital cost of continuous-casting installations depends very much on the individual circumstances, in exactly the same way as does the cost of rolling mills. As one example, however, a study was made recently of the relative costs for the production of 500 000 tons/year of heavy strip mill slabs in conjunction with pneumatic steelmaking, which

* G. Hoyle and E. Ineson: *JISI*, 1959, vol. 191, Jan., pp. 44–55.

TABLE 2—Bend test results on bar similar to that used for Table 1 (O Q 1 280° C, D T 560° C)

Property	Outer band	Inner band
V H N	875	847
Limit of proportionality, tons in²	112	112
Apparent breaking stress, tons in²	165	186
† True breaking stress, tons in²	131	139
† L/θ	0·66	0·66
† L/ϕ	0·90	0·80
† Ductility factor Y	32	47
† Quality factor Q	17·5	23·5

† Empirical values derived by Hoyle and Ineson

showed that the cost of the slab mill installation would be 2½ times as great as the cost of the equivalent in continuous casting. As the output increases towards the saturation point for the largest type of slabbing mill the investment costs come closer together, but even at this very high level of production, there still remains a real advantage on production costs of something like (in this instance) £2/ton. At lower tonnages, where the high-capacity slabbing mill is not necessarily used to the maximum, the advantage is considerably greater.

Mr **W. A. Johnson** (The United Steel Companies Ltd) wrote: The development of effective continuous-casting practice seems to have been built up in the best tradition of scientific investigation, leading to systems which work and work well. But a general look at the principles, artifices, and plant adopted suggest elements of empiricism which are dealt with somewhat arbitrarily in the papers presented.

Amongst these features are:

1. The use of both stopper and lip pouring
2. The use of both fabricated and cast moulds
3. The voluntary or involuntary vertical oscillation of the mould and the use of static moulds
4. The use of a rather small mean temperature rise in the mould cooling water, even allowing for a higher temperature gradient in the upper part of the mould
5. The somewhat complicated method used at Barrow to support the secondary water-cooling sprays.

Dr **Waters** wrote in reply: We agree with Mr Hoyle that the bands in continuously cast high-speed steel bars are not composed of carbides alone. This interpretation of the photograph (Fig. 32) was not intended but nevertheless heavy carbide segregates are associated with this banding. In practice, it has been found that toolmakers will not accept material containing this banded type of structure. It is considered that this decision would not be altered even if it could be proved that no significant difference in properties, as revealed by mechanical testing, existed between the central and outer portions of the material.

Many hundreds of samples of continuously cast high-speed steel have been examined by ourselves and also by other steelmakers and in almost every case the centre third of the bars contained heavy carbide segregates and were considered unsatisfactory.

It must be pointed out that no work was done on extrusion as it was considered that this would not greatly improve the quality. The sample examined by Mr Hoyle may, however, have had the segregates partially broken and dispersed by the extrusion process.

Discussion on creep

This discussion was based on the following papers (dates of publication in the *Journal* are given in parenthesis):

"A new approach to the problem of creep," by *J. Glen* (1958, vol. 189, Aug., pp. 333–343).

"The effect of alloying elements on creep behaviour," by *J. Glen* (1958, vol. 190, Oct., pp. 115–135).

"Ductility in high-temperature rupture tests," by *J. Glen* (1958, vol. 190, Sept., pp. 30–39).

"Recent developments in creep testing by the cantilever bending method," by *G. T. Harris, H. C. Child, A. B. Collier,* and *C. F. West* (1959, vol. 190, Oct., pp. 136–143).

Dr **J. Glen** (Colvilles Ltd) presented his papers, and Mr **H. C. Child** (Wm Jessop and Sons Ltd) presented the paper by himself and his co-authors.

Mr **L. E. Benson** (Metropolitan Vickers Electrical Co. Ltd): In these and other papers Dr Glen has acknowledged the rightness of certain ideas about the nature of creep first put forward by Dr Bailey. As one of those fairly closely associated with Dr Bailey's work, I am certain that if he were here he would be the first to express appreciation of Dr Glen's work and, in particular, of these papers.

An impressive feature is the comprehensive picture presented of the factors governing creep from the beginning of a creep curve to the end, starting with strain hardening and finishing with rupture and ductility at rupture. The main theme is what happens in between the beginning of the test and the final tertiary part of the curve, and in dealing with this Dr Glen deals with the effects produced by 10–12 elements in steel which operate in different ways to affect the resistance of the lattice to creep. I do not recollect such a comprehensive and rational picture having been attempted before and I believe that there will be general agreement about the merit of this conception.

Dr Glen has reminded us that in the initial stages of a creep test the slowing up in the creep rate is due to strain hardening. He takes this for granted, but it may strengthen the case to point out that the dislocations which stiffen the lattice can be introduced by cold working before the creep test starts.

Figure *A* shows a practical application concerning austenitic steel bolts. When austenitic steel constructions were first used for gas-turbine work and other high-temperature applications it was found that along with their low yield point the bolts had an inconveniently high initial creep strain. The curves in the figure show the results of some creep tests on 18–11–1-type bar. The two top curves may be taken to represent the variation found in normal commercial material and the lower curves relate to test pieces which were stretched cold by different small amounts before starting the creep test. The improvement in the creep rate right from the start indicates the effect of strain hardening. Also there is a marked irregularity in the top curves in the later stages of the test, which may well be due to strain-age hardening as postulated by Dr Glen.

At the other end of the creep curve he mentions the possible damage effect on creep rate of grain growth on recrystallization. His paper is not about this and it might not often be a factor in steel under normal service conditions, but we must bear it in mind as a possibility.

The most important part of Dr Glen's work concerns what goes on between strain hardening and the tertiary part of the creep curve. In a paper to the World Power Conference in 1929 R. W. Bailey showed that structural changes visible under the microscope occur in superheater tubes at quite moderate service temperatures. I mention this because it is an interesting piece of history. In 1932 in a paper to the Institution of Mechanical Engineers he reported further that structural changes in a number of steels had been studied and found to obey the now familiar law for similar conditions in which log time varies as the inverse of the absolute temperature. In 1954 Dorn and his associates related strain as well as structure to time and temperature in the same way and now Dr Glen has taken us a most important step further by indicating by a most ingenious technique that existing creep data, even creep data produced in different laboratories, contains strong evidence of the occurrence of structural effects in commercial steels and of the way in which they determine the creep rate.

This work raises many points for discussion. For example, there is the advantage of Dr Glen's method of plotting as an aid to extrapolation. It might be thought to have some limitations and difficulties, and it is possible that the planning of creep tests will have to be materially altered from that to which we have been accustomed in the past if we are to get the most value out of them.

There are some items which have not been dealt with or have been dealt with insufficiently, one of which is the effect of carbon. For instance, the value of rupture tests as a guide to working stresses may be discussed, and it may be possible to corroborate some of Dr Glen's rather circumstantial evidence.

These and other questions deserve the fullest discussion but I want to make two further points. The first is that these papers should give a stimulus to the study of structural changes. If our present alloys are to be improved, the best chance of doing this is through a better understanding of what goes on inside the steel. In the UK we have a good start in this field, and we should not lose it.

To underline the importance of this I should like to refer to two more figures. Figure *B* shows a carbon replica structure of a molybdenum–vanadium steel which has been normalized from 1 000° C and then tempered for 16 h at 700° C, i.e. the sort of drastic tempering treatment that may be given in practice to reduce heat treatment stresses. The structure is a more or less uniform distribution of carbides which, at higher magnification and when properly prepared, can be seen to have a crystalline habit.

Figure *C* shows the same material after a further treatment simulating the effect of a normal lifetime at 1 050° F, which is the highest steam inlet temperature in power station work in the UK today. There is an obvious change, for the larger particles have grown at the expense of the smaller. In the face of such a change it does not require imagination to realise that changes may be going on, on an even smaller scale, such that creep behaviour may be affected profoundly.

The other point I want to make is that if Dr Glen's conception is correct, it must apply on a much wider basis than just to the steels about which we have been talking. Dr Glen himself touches on this, but it is

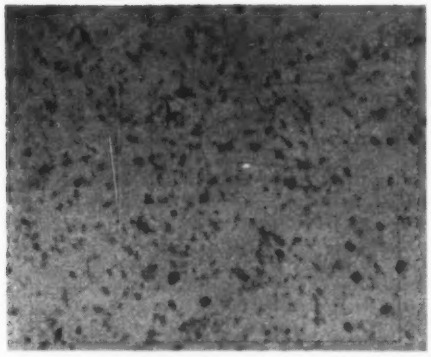


Fig. *A*—Structure of Mo–V steel after normalizing from 1 000 C and tempering at 700 C, electron micrograph × 32 000

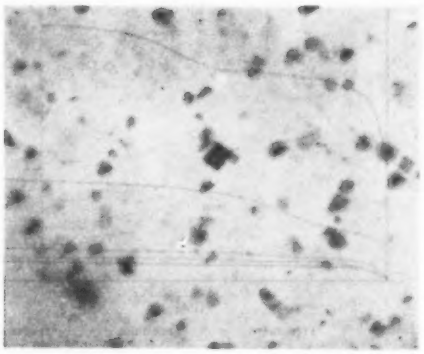


Fig. *B*—As Fig. *A*, but showing change in structure owing to further treatment to simulate 100 000 h at 565° C, electron micrograph × 27 000

interesting to see the results of trying a Glen plot on two zirconium alloys (Fig. *D*). These alloys were referred to in a paper at the recent Geneva Conference. One is a zirconium–copper alloy and the other is the same alloy with molybdenum added. The sensitivity of the equipment is such that the humps in the curve are genuine. It looks as though Dr Glen's conception applies to these alloys. Furthermore, the addition of molybdenum confers a considerable advantage in reducing the creep rate, and this appears to result not from the addition of new humps in the curve so much as from the 'concertina-ing' of the whole curve vertically. Perhaps Dr Glen will be good enough to comment on this.

Mr **J. D. Murray** (The United Steel Companies Ltd): The great advantage claimed for Dr Glen's new method is that it shows transitions or bumps in the creep curve which are presumably due to metallurgical phenomena. Consequently, since any extrapolation by such methods takes into account metallurgical changes it must give one a great deal of confidence.

We have all known for some time that these metallurgical changes take place, but I am in some doubt about the explanation given for them by Dr Glen. The phrases which he uses, e.g. strain-induced precipitation and strain-age hardening, which in the papers are one and the same thing, seem to be used to cover a multitude of sins. If it is meant to be a new phenomenon which occurs only in a mechanically strained lattice and which does not occur without the presence of mechanical strain, then there is still a large amount of work ahead of us. This can hardly be so in view of the large amount of work which has been done on the effects of cold and warm work. It seems to be a well known phenomenon which must be fitted into the picture.

What are the possibilities? First, there is the effect of atmospheres; second, the coherency strains associated with the precipitation of carbides and nitrides; third, the precipitated particles themselves; fourth, spheroidization, and fifth, recovery and recrystallization.

From the form of the log strain/log creep rate curve the whole of a process, or, rather, the repetition of a certain process, occurs, i.e. it starts and finishes during testing. Consequently, precipitated particles and spheroidization can be eliminated. It is almost certainly not recovery and recrystallization, although it is true that transitions due to such phenomena have been noted in the creep testing of pure metals, and it could also be possible that recovery from the initial loading strain and the strain hardening incurred in the primary stage of creep could cause one transition, as already shown by Mr Benson.

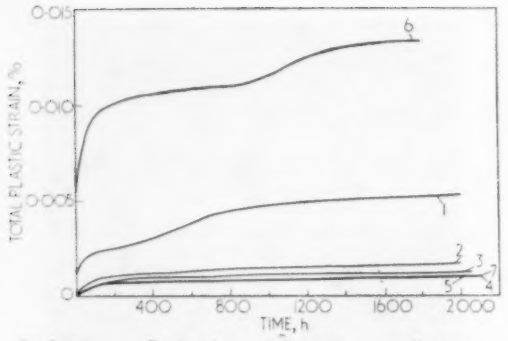
We are therefore left with atmosphere effects and coherency strains associated with precipitation. Dr Glen has suggested that the effects are due to some form of coherent precipitate of carbide or nitride in dislocations rather than the formation of atmospheres as postulated by Cottrell. At United Steels we have some evidence that in fact both phenomena are involved. First, in aluminium-killed carbon steel we have isolated the effect of nitrogen in creep tests at 450° C and have had the opportunity of examining many of the specimens involved, some of which were on test for 40 000 h, with the electron microscope. In no instance did we find any evidence of a nitride precipitate other than the aluminium nitride present in the steel in the normalized condition. Consequently, we have convinced ourselves that in the circumstances the nitrogen effect is one of atmospheres.

Second, from creep tests on 0·5%Mo steel and associated electron microscopical examination we have shown that creep strength increases up to an optimum, that at this optimum coherency strains are at a maximum and that beyond the optimum the Mo_2C loses coherency and starts to over-age. There is a great similarity between atmospheres effects and coherency strains, although, obviously, both are distinctly different. With atmospheres we have small interstitial atoms clustering around a dislocation and restricting its movement, while with coherency strains we once more have a clustering but it is of larger substitutional atoms around carbon atoms, the latter being presumably at dislocations. With respect to the latter effect, in such regions the metal lattice is greatly strained, and it is this strain which effectively blocks the movement of dislocations.

Combining these two phenomena, atmospheres and coherency strains associated with precipitation, it is possible to explain all the transitions which Dr Glen shows in those steels tested in the normalized or solution-treated condition. However, it is the tests on the steels which have been tempered which puzzle us, and I should like to ask Dr Glen one or two questions.

In Fig. 5 of the first paper he shows the results of a carbon–manganese steel tested after normalizing and tempering for 2 h at 650° C. What precipitation does he suggest are formed at 500° C and below during creep testing, bearing in mind that almost certainly 2 h at 650° C will have caused the Fe_3C to start to spheroidize. In addition, the manganese will have diffused to the Fe_3C. It is rare, it should be noted, to find manganese carbide; what usually happens is that manganese replaces some of the iron in Fe_3C. We have evidence to show that in a similar type of steel the nitrogen effect is destroyed by tempering for 1 h at 660° C.

Similarly, what precipitates form in the Mo–V steels



Specimen	Treatment	Specimen	Treatment
1	none	5	6% cold work
2	1% cold work	6	16 h at 850° C
3	2% ,, ,,	7	16 h at 650° C
4	3% ,, ,,		and 6% cold work

Fig. C—Heat treatment and cold-work effect on FCB(T) bar, all tests at 600° C and 9 tons/in²

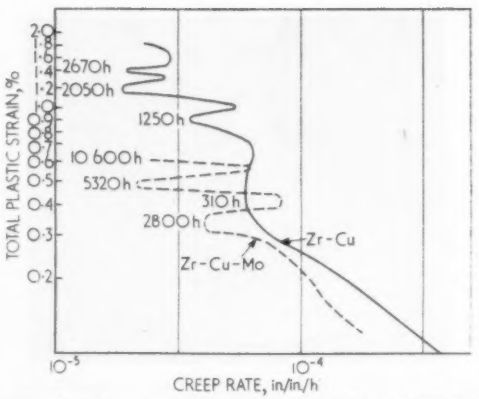


Fig. D—Strain rate curves for two Zr alloys, 6 000 lb in² at 450° C

tested after normalizing and tempering for 5 h at 690° C? Such a tempering treatment is just beyond the secondary hardening peak, and one would imagine that the transition due to vanadium is in progress as the creep test begins.

A similar question arises on the ½%Mo–V–Ti steel. The author states that a treatment of 1 150° C takes TiC and TiN into solution. Our experience is that TiN is not taken into solid solution by heat treatment and we have evidence that it persists up to solidus temperatures. How did he show that TiN was in solution?

Turning to the paper "Ductility in high-temperature rupture tests," I agree with Dr Glen that incrystalline failure in ferritic steels is governed more by metallurgical changes taking place in the body of the grain than changes at the grain boundary region. This does not imply that the boundary is not of considerable importance. The same cannot be said, however, of austenitic steels, since in such steels precipitation takes place mainly at grain boundaries and not within the grain.

In the paper much of the hypothesis is based on the work of Dorn, who suggested that the same creep law applies to grain boundary creep as to total creep of pure metal. If my memory serves me correctly, Dorn's work was done on high-purity aluminium and he claimed this behaviour only at temperatures greater than half the melting point. He pointed out that it would not necessarily follow at temperatures less than half the melting point, which are conditions more analagous to those which we face in steel.

Because of this, and also on other metallurgical grounds, it is difficult to agree with Dr Glen that it is reasonable to make the assumption that the same creep law applies to both total creep and grain boundary creep in complex alloys. One might expect a similarity in the creep curves of total creep and boundary creep if precipitation is confined to grain boundary regions, but it is difficult to imagine such a similarity if precipitation is mainly taking place in the body of the grain.

The postulation of a normal and an embrittling mechanism of failure is partly, if not wholly, correct, but one is tempted to say that it is right for the wrong reasons. The evidence given for ½%Mo and 1%Cr–Mo in the papers is much too speculative. If Figs. 7 and 8 are examined closely it will be seen that in no case is the transition attributed to molybdenum shown up by experimental results. There is no doubt that the low ductility of ½%Mo steel is associated with the precipitation of Mo_2C within the ferrite grains, and I suggest that the minimum in the ductility occurs when coherency strains are at their maximum. Consequently, is not this a simple case of the grain being strengthened relative to the boundary region? As the Mo_2C loses coherency and migrates to the boundary region, picking up more alloy to form M_6C, the strength of the grain decreases and, provided that severe strain damage has not occurred, the balance between boundary and grain is restored and ductility increases.

The suggestion that Cr delays the Mo transition could not be supported from the metallurgical evidence of the effect of tempering this steel, since it has been shown in a number of steels of this type that the effect of Cr is to tend to accelerate the secondary hardening due to molybdenum.

Would Dr Glen like to comment on the fact that in the plot of stress/elongation the stresses at which the minimum ductilities are obtained approximate very closely to the respective 0·1% proof stresses at the temperatures in question. Consequently, all the specimens will have received a type of warm working during the loading operation. A well known effect of warm working the alloys is to lower the rupture ductility.

Dr **J. Heslop** (Mond Nickel Co. Ltd): Dr Glen has shown in his previously published work that strain ageing occurs in steel during tensile testing. He has now presented considerable evidence for strain ageing in steels during creep and has also suggested that embrittlement is due to this strain ageing occurring at different times at grain boundaries and within the grains.

I am concerned with alloys of the Nimonic series, some data for which Dr Glen has analysed in his papers. With these complex alloys, although there is metallographic evidence that creep deformation accelerates and in some instances modifies precipitation, there is no evidence in support of conventional strain ageing. The photomicrograph of a Nimonic 90 creep specimen tested to rupture at 650° C shows a particular type of precipitation occurring at grain corners (Fig. *E*). This does not occur merely by ageing at 650° C and is evidence of strain-induced precipitation.

Creep tests discontinued after being subject to the same stress and temperature conditions for different times have been tensile tested at room temperature to determine the amount of embrittlement occurring during creep. Room temperature was chosen for the tensile tests to eliminate the possibility of any further precipitation. If strain ageing occurred during creep it would be expected to produce an increase in proof stress and a decrease in elongation at room temperature. In fact, it was observed that although elongation decreased with creep strain, the proof stress and UTS remained essentially constant.

From these and other studies we conclude that, although precipitation is accelerated and modified by strain, the conventional strain ageing observed in steels does not seem to occur in the Nimonic alloys.

In connection with intergranular failure, which is dealt with by Dr Glen in his final paper, our experience is that the mechanism of fracture is primarily influenced by the conditions at the grain boundary region. The types of intergranular rupture observed in Nimonic alloys under creep conditions have been classified into three groups according to the microcracking adjacent to the main fracture:

(i) Where no microcracking is visible
(ii) Where microcracking occurs at grain junctions and adjacent grains are pulled apart
(iii) Where microcracking occurs by the formation and subsequent growth of cavities at grain boundaries.

The fracture mechanism appears to be controlled by creep ductility; the observed elongation to fracture increases in the order described above. In the Nimonic series of alloys elongation to fracture can be varied markedly by variations of heat treatment which produce observable metallographic differences in the grain boundary regions. The distribution of fracture types for the normal condition of heat treatment (i.e. 8 h/1 080° C — 16 h/700° C) is shown in Figure *F* and it can be seen that the most ductile type of rupture (iii) occurs under test conditions involving low stresses and high temperatures. These conditions also allow further precipitation at grain boundaries during testing. At lower temperatures or higher stresses the precipitation during creep is reduced and the strain rate is increased. The grain boundaries are consequently less able to relax stress concentration and the less ductile type of rupture (ii) occurs. Intragranular precipitation during creep further reduces the ability of boundaries to relax stress concentrations and the most brittle conditions is obtained.

When heat treatment is chosen deliberately to suppress precipitation at grain boundaries the most brittle type of failure occurs over a wider range of conditions. From this evidence it is concluded that the grain boundary regions and the effects of precipitation occurring there are the factors which primarily govern the type of intergranular failure.

Dr **G. Bandel** (Bochumer Verein): I have only one question to ask, and that is whether Dr Glen has found negative creep. I found it for example on short-time creep tests at 500° C of steels with 0·11%C, 0·56%Mo, and 0·55%Ti after normalizing without tempering.[1] I also found it in austenitic steels similar to nimonic alloys, solution treated only. Perhaps for commercial steels this is not important, but it might be important in investigating the causes of the transition points, mentioned in the papers.

Mr **H. H. Smith** (English Electric Co. Ltd): It is at once evident from Dr Glen's conclusions that there is no short cut to the determination of the long term behaviour of metals under creep conditions at elevated temperatures. However, what are Dr Glen's views on the usefulness of short-time tests as acceptance criteria for metals to be used in long-time service? It is frequently necessary to undertake relatively short-time rupture tests to obtain information about longer term behaviour. In this case it is usual to adopt a series of relatively high stresses at a testing temperature somewhat above the operating temperature. Is it more reliable to test under high stress at the working temperature, or under the working stress at a higher temperature?

Also, for how long should such tests be carried out to give reliable extrapolation to 10 000 h: or is the rupture test simply a guide to the ductility of the steel?

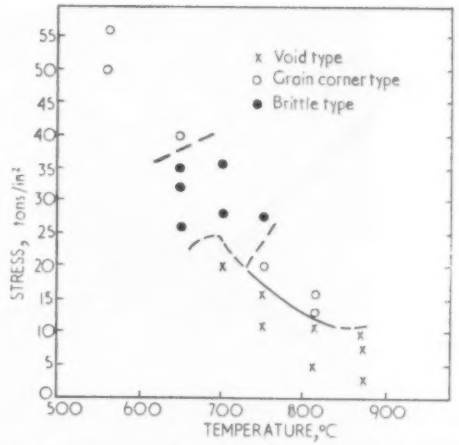


Fig. *E*—Rupture mechanism as a function of stress and temperature for Nimonic 80A

Dr **R. G. Baker** (University of Cambridge): It is now generally accepted that the phenomenon of strain ageing at low temperatures can be explained by the segregation of solute atoms to stationary dislocations, as suggested by Cottrell and his co-workers.[2, 3] Cottrell[4] has also considered the case of moving dislocations somewhat as follows. At temperatures sufficiently high for atomic migration to take place fairly readily, a solute atom may move while interacting with a dislocation. If this happens, the atom will jump preferentially into a position where it has a lower energy of interaction, i.e. towards the dislocation. The excess energy which it loses by this movement will then be dissipated in lattice vibrations. If it is to go on moving, the extra energy to compensate for this loss must either be supplied from the kinetic energy of the dislocations themselves or must be found by raising the external stress. This means that unless the external stress is raised, the velocity of the dislocations and hence the strain rate will also decrease.

The solute atoms will have their maximum retarding effect when their mobility is such that the atmosphere can just be dragged along at the velocity of the dislocation. If their mobility is much less than this, solute atoms will be unable to catch up with the dislocations to form an atmosphere and, if it is much greater, they will be able to migrate with the dislocation without seriously impeding it. It is not necessary for the solute atoms to be in interstitial positions to segregate around a dislocation with decrease of interaction energy. On one side of an edge dislocation the lattice is compressed and on the other side it is extended, and one will tend to get the larger atoms segregating to one side and the smaller atoms to the other.

During the progress of a creep test the diffusion of the solute atoms is continually speeded up because of the creation of lattice defects during straining. On the other hand, the velocity of the dislocations is becoming steadily smaller as it becomes more difficult for them to move through the lattice. Ultimately one reaches a stage when the most mobile of the solute atom types present can form atmospheres, which will then be dragged along by the dislocations. The extra energy required to move them, if the stress remains constant, will be largely drawn from the kinetic energy of the dislocations themselves, and the strain rate, and hence the creep rate, will decrease. This might explain the strain-ageing effect.

As the mobility of the solute atoms increases still

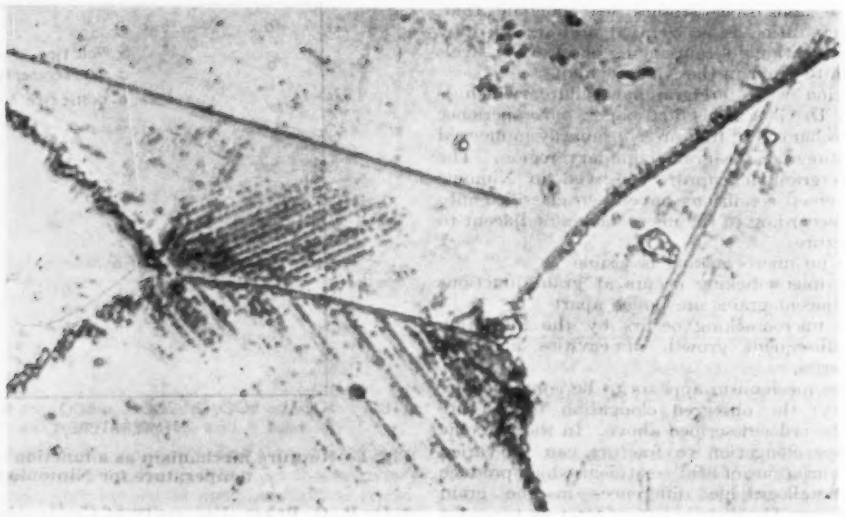


Fig. *F*—Nimonic 90 tested to rupture at 23 tons in^2 and 650° C × 2 000

further, the atmospheres become progressively less effective in impeding the dislocations. This may be the explanation of the apparent over-ageing effect observed in some creep tests. Alternatively, the solute atoms may congregate in sufficient numbers for precipitates to be nucleated along the dislocation lines. The nuclei would be expected to grow until they had absorbed the surrounding solute atoms. The number of solute atoms might be considerable. It has been calculated that up to about 40 solute atoms per atom length of the dislocation line may congregate. The nuclei would continue to increase in size by growth at the expense of each other, and this would also be expected to give rise to an over-ageing effect.

Thus strain ageing in creep may be explained in terms of the formation of atmospheres, whether or not precipitation results. We might expect all the solute elements in turn to give rise to strain-ageing effects in order of their mobilities. It is difficult in some cases to envisage how and in what form precipitation could occur, as, for example, in the case of copper, especially if n any of the other solute atoms had already been tied up in compounds which had formed in earlier strain-ageing operations. However, in individual cases it should be possible by the use of extraction replicas and thin film techniques to discover whether or not segregation is being followed by precipitation.

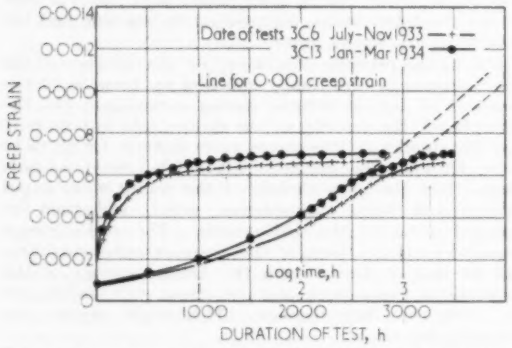


Fig. *G*—Cast steel 3C6, 0·3%C, stress 6 tons/in^2 at 440° C

It might be expected that the onset of strain ageing would occur at lower values of total strain as the temperature was increased, if it is assumed that lattice defects are increasing during creep and increasing the diffusion rate of the solute atoms.

The experimental results show quite clearly that the reverse was the case. The reason for this is probably that two other factors must also be taken into consideration. The first of these is that the rate of annihilation of point defects will be greater the higher the temperature. Second, the strain rate will be higher at higher temperatures, and this means that the dislocations will be moving at a much greater velocity. Thus, an atmosphere is required to migrate much faster to keep up with a dislocation and exert its maximum restraining force.

I should like to comment briefly on the mechanism of fracture suggested by Dr Glen. One of the assumptions was that the concentration of solute in his model at the grain boundaries was less than in the interior of the grain. This may not always be so, for example the ideas of equilibrium segregation to the boundary suggest that the reverse is more likely to be true. It is possible to

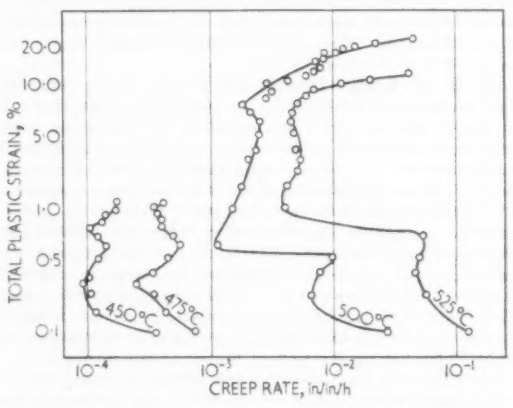


Fig. *H*—Strain rate curves for Zircaloy 2, 6 000 lb/in^2

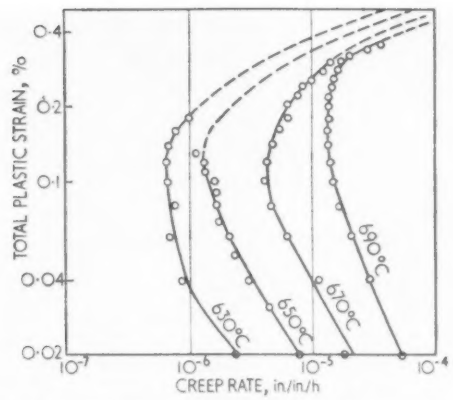


Fig. *I*—Strain rate curves for Mo–V steel steam pipe, 3 tons/in²

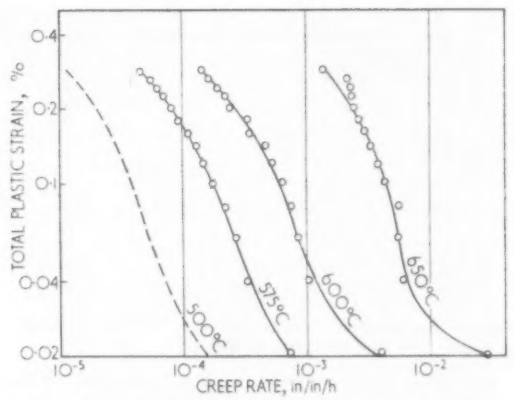


Fig. *J*—Strain rate curves for $2\frac{1}{4}$%Cr–1%Mo steel, 3 tons/in²

explain the results qualitatively on a simpler model, along the lines suggested by Mr Murray. If grain-boundary sliding can take place more readily than grain deformation, stress concentrations will be built up at the boundaries and this may result in the initiation of cracks. It follows that if the grains increase in strength, the stress concentrations will be built up more readily and cracks are more likely to be initiated intergranularly. When over-ageing takes place the stress concentrations will be dissipated in plastic strain as the grains become weaker; and if fracture has not already occurred one might expect the material to go on quite happily until another effect of this kind takes place.

The picture presented by this argument is probably over-simplified, as was pointed out by Dr Heslop. Very little is known about the effect of segregating elements on the properties of the boundaries. During strain ageing the alloy concentration at the boundary is undoubtedly affected, and this might affect in turn the ease with which cracks were nucleated and propagated at the boundaries. It is premature to attempt a detailed explanation until the effects of the elements on the boundary properties have received more study.

Mr **J. F. Mercer** (Murex Welding Processes Ltd): As a comparative newcomer in the field of creep testing I have been somewhat disturbed by the amount of work that is based on mathematical extrapolation of creep data. I do not think that anyone who has had to contend with test data for steel on a cast-to-cast basis would feel very happy about treating steel as a homogenous material easily expressed in terms of equations and formula. It is sometimes difficult to make this point clear and Dr Glen's papers should provide much useful guidance in this matter.

Would Dr Glen consider it feasible to predict, in the light of his theories on strain ageing, how much precipitation will be necessary to prevent failure in any particular time under given conditions? Does he think it possible to manipulate the 'bumps' so as to produce a steel which will be ideal for the service envisaged. It might be practicable, now that the transition in the creep curve can be shown so clearly, to design creep-resisting steels on this basis rather than by empirically alloying the steel. Dr Glen's views on this point would be appreciated.

Mr **J. H. M. Draper** (Metropolitan-Vickers Electrical Co. Ltd): Dr Glen has produced a very fine model, giving convincing explanations of the various creep phenomena. I feel that one cannot criticize the work in general terms, and one could not in any case criticize it in any detail without countering with a lot of data.

Dr Glen has elaborated on the late Dr R. W. Bailey's ideas of thermal influence in creep, and suggests that his own analysis includes the strain age hardening effect which Dr Bailey's pre-heat treatment did not. Nevertheless, I am not yet clear in my mind as to the magnitude of the effect on creep of the strain-age hardening factor, relative to that of age hardening or precipitation hardening, which is unaffected by strain.

Figure *G* is of historic interest. Dr Bailey put it forward in 1935[5] as an example of age hardening. When the creep curve is plotted with log time as the abscissa it reveals what appears to be a distinct transition. I do not put that forward as an alternative way of plotting because a transition has to be considerable for it to show up in this way.

I commend Dr Glen for the enormous amount of work he has put into the papers. Recently, members of our creep laboratory have had an example of the effort of analysing just a few curves. During this time we have discovered, amongst other things, that the strain/rate curve is extremely sensitive to slight changes in the shape of the creep curve, and that the tests have to be very accurate. With tests of medium sensitivity one has to use a lot of discretion in plotting the strain/rate curve. In the limited time available we tried to plot strain/rate curves of materials which did not appear in Dr Glen's papers. Figure *H* shows strain/rate curves for tests on one of the modern alloys used as a structural material in a reactor, Zircaloy-2.[6] It is an interesting family of curves, not easy to interpret, but serving to show that we ought to have carried out some of the tests for longer times. It should be noted that these were medium sensitivity tests.

Looking through the papers we found no strain/rate plots of $2\frac{1}{4}$%Cr–1%Mo steel and only one for normal Mo–V steel. We therefore tried to plot some for these two steels, and Fig. *I* shows strain/rate plots for Mo–V steel. We did not feel justified in plotting strains lower than those shown, and the curves do not show any transitions. Either the transitions must have taken place below 0·02% or the good creep resistance must be in the initial resistance of the material.

As a contrast, Fig. *J* shows strain/rate plots on the same scale for $2\frac{1}{4}$%Cr–1%Mo steel. Whereas the previous curves revealed no transitions and showed the points at which the creep rates start to accelerate, in this case there appear to be transitions at about 0·05% strain, and these transitions continue right up to 0·3% strain, the limit of the tests.

Fig. *K*—Strain rate curves for 2¼%Cr–1%Mo steel, 3 tons/in² at 650° C

We also attempted to extrapolate using Glen's suggested method of constructing a line to represent another temperature and by integration we obtained a point on the temperature/log time curve which agreed reasonably well with the normal extrapolation.

In Fig. *K* we have taken the curve for Mo–V steel with the corresponding curve for the 2¼%Cr–1%Mo steel from the previous two figures. The curves show up very well the superiority of the former steel at small strains, hence the reason for its preference as a material for such applications as steam turbines. As the strain increases, however, the Mo–V steel is worsening, and it goes into the tertiary stage at about 0·15%. The Cr–Mo steel, on the other hand, is improving and showing no signs of increasing creep rate. While the tests do not go far enough to prove much, the strain/rate curves indicate the reason for the controversy which exists regarding the relative merits of the two materials for applications in which large strains are permissible.

Mr **J. M. Arrowsmith** (University of Cambridge): At Cambridge we have been making a study by electron microscope of phases precipitated from G18B, an austenitic steel based on iron, nickel, chromium, and cobalt, with small additions of molybdenum, tungsten and niobium. Direct evidence has been obtained of precipitation of dislocations both during ageing and during high temperature straining. On solution treating for 3 min at 1 300° C the molybdenum, tungsten, and some of the niobium goes into solution and segregates to vacant sites of the lattice. Figures *L* and *M* show the precipitation which occurs on ageing 20 h at 700° C. This takes place along the dislocation lines. The dislocations either originate from the forging treatment and have survived the subsequent heat treatment or, alternatively, they have been formed in the specimen by air cooling stresses.

Figure *N* shows in detail a dislocation on which has precipitated large cubes of $M_{23}C_6$ and much finer crystals of niobium carbide.

In the paper 'Ductility in high-temperature rupture tests' Dr Glen reproduces in Fig. 10 a strain rate graph for the solution treated G18B creep tested at 700° C. Two transitions in creep rate occur. Could this be explained by the precipitation in the dislocations of first the cubic $M_{23}C_6$ and later the niobium carbide?

When G18B after solution treatment is deformed at 700° C many dislocations are generated. This treatment is known as warm working. Niobium carbide crystals are independently nucleated along the length of these dislocations and grow to about 25 Å across. The whole grain is filled with this precipitation (Fig. *O*).

The superior creep resistance at 700° C of the warm worked material over the solution-treated material can

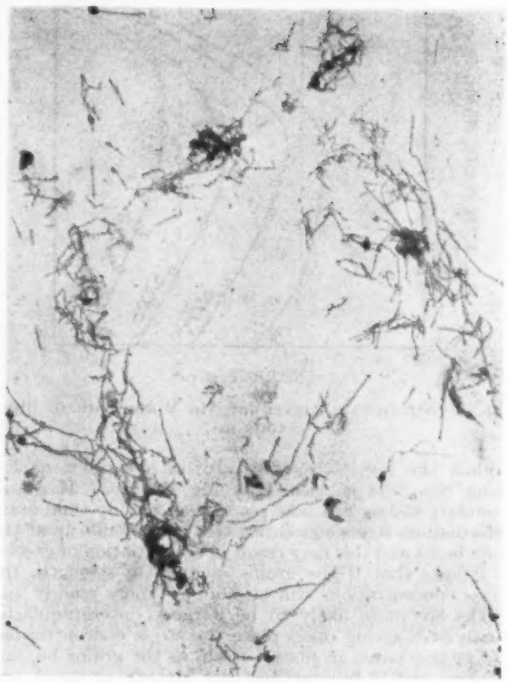


Fig. *L*—Specimen G18B, solution-treated and aged 20 h at 700° C; carbon extraction replica (photograph reduced × ⅔) × 4 700

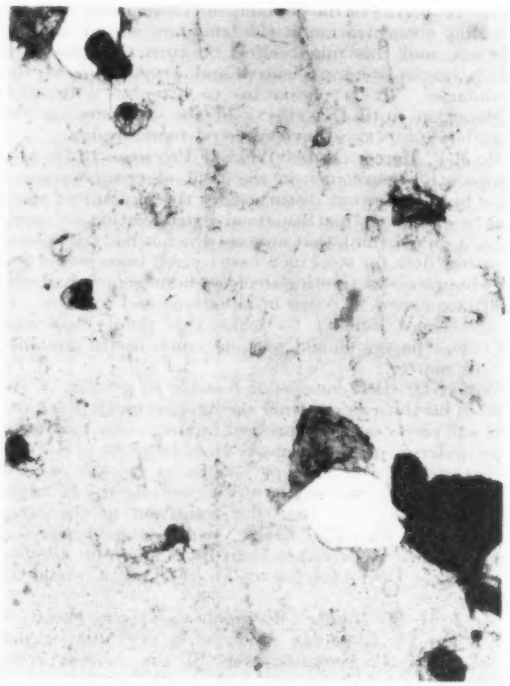


Fig. *M*—As Fig. *L* × 25 000

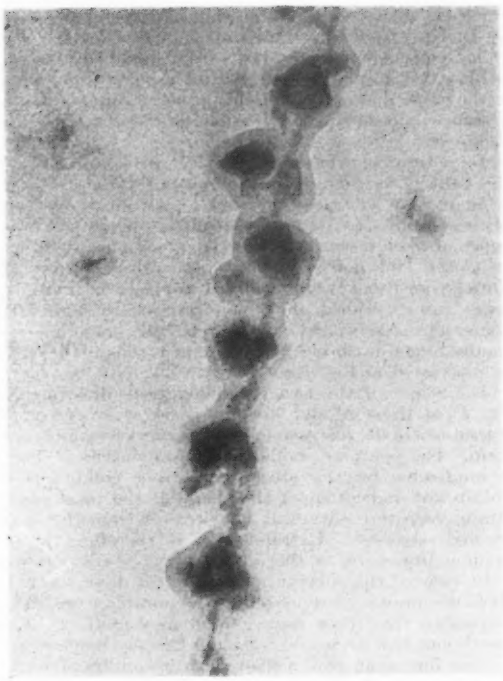


Fig. *N*—As Fig. *L* × 160 000

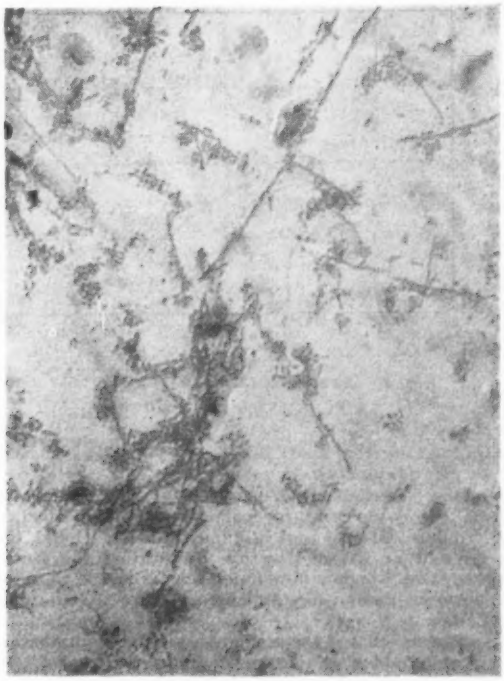


Fig. *O*—Specimen G18B, solution-treated and warm worked 10% at 700° C; carbon extraction replica × 160 000

be explained in terms of the large increase in initial creep resistance caused by a high density of dislocations and fine precipitates locking them. In addition to the high resistance of grains to further deformation by creep, the subsequent coarsening of the carbide will be retarded by virtue of the reduced concentration of niobium left in solution.

Mr **H. W. Kirkby** (Brown Firth Research Laboratories): I think that most of us will agree that the transitions which Dr Glen has shown occur in his creep curves, and that they are brought out very much more quickly by his method of plotting. At least some of them must bear some relationship to strain-ageing mechanisms but my main difficulty is deciding whether a transition is related to the material or whether it is connected with the creep-testing procedure.

Some of the transitions in Dr Glen's curves are very clearly defined and cannot be denied but others are not as clearly defined and I find that I have to stretch my imagination a little to convince myself that they are transitions. These comments apply particularly where there are a number of these 'wiggles.' In scanning a number of steels, mostly complex, we have noted these changes in slope, but I must confess that some of them are most peculiar. This may be due to testing or it may be genuine. I should like to discuss this further with Dr Glen.

I agree that the present methods of extrapolation are not very satisfactory, but I am not entirely convinced that his method of deriving creep curves from strain rate data for extrapolation purposes will provide information of sufficient accuracy. He did not demonstrate a 'before and after' series, but no doubt that was due to lack of information.

We have a few series of different stress and temperatures which extend to 20 000–25 000 h and which we will gladly place at his disposal. I should like to give him half a series and ask him to predict the other half.

I think that Fig. 12 of the paper 'Ductility in high-temperature rupture tests' would not be as simple as shown if we had more data. We have tested 1%Cr–Mo steel, and our ductility plot on the basis of Fig. 12 is very similar to the C–Mo steel. In the same way, turning to a more complex steel, namely stainless iron with molybdenum, in rupture tests at 550° C and up to 30 000 h, this steel shows no sign of any 'normal mechanism.' If you plot it on the basis of Fig. 12 it is practically a horizontal line.

It may be that there is some explanation, but I feel very unhappy about Fig. 12 and I think that there is a certain amount of luck about it. With varied treatments one can get all sorts of shapes and slopes, although these can be discussed later.

On titanium carbide and nitride, I think the answer to Mr Murray is that the bulk of the carbide and nitride in many steels never goes into solution. In the case of the carbide in a number of steels it is in the eutectic phase. The nitride often comes out before the steel freezes, but a small proportion goes into solution; this varies with the steel, and it may be 50–50 or 75–25. I think one can get titanium carbide in solution.

Mr **R. P. Kent** (C. A. Parsons and Co. Ltd): I will deal only with the extrapolative aspects of the papers. It seems to me that if these transitions are genuine, which can probably be proved by an accumulation of data, this is a very fundamental way of extrapolating data. It is certainly more fundamentally sound than extrapolating short-duration tests on a log-linear basis.

I should like to draw attention to Fig. 9 in the paper 'A new approach to the problem of creep.' Here we have iso-strain curves for a 1·2%Mn steel which indicate bending. This reveals the hazards of straight-line

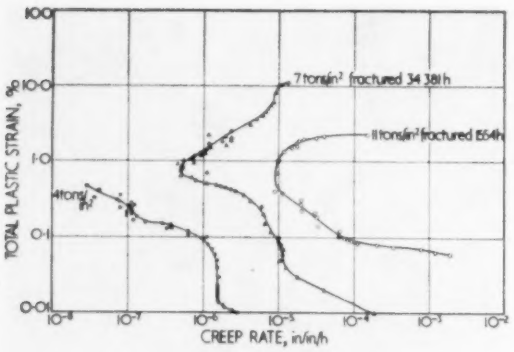


Fig. *P*— Creep data for G 18B at 700° C

extrapolation from short duration tests, and brings up a very important aspect of creep testing, i.e. the provision of long-time data, particularly scatter data, for steels. It is essential, for particular types of steel, to accumulate data, to bear in mind the scatter band, and to present an average picture by reading the scatter bands. One wonders how much scatter is due to extrapolative errors and how much to genuine metallurgical variables.

It appears that if the scatter is due to genuine metallurgical variables, an average set of data for the 1·2%Mn steel is not enough because the iso-strain curves are so close together at service temperatures that data will have to be established for each cast of this steel. Dr Glen mentioned the probable errors involved in temperature for 0·1% strain in 100 000 h and said they were not great: the figure given was 427° C ± 10° C. It may also be considered in terms of plastic strain. At the extremes of these temperatures, the plastic strain for 100 000 h varies from 0·03 to 1·0%, which is a considerable difference.

This probably means that if the metallurgical variables are the cause of scatter the whole of the strain curves for this particular steel could be displaced either one way or the other; using 427° C, you could get 1% creep strain, or at the other extreme 0·03%. This seems to place great emphasis on reliable methods of extrapolation. It will eliminate one source of error in the scatter band.

Mr **H. C. Child**: I should like to confine my comments to two aspects of Dr Glen's paper which have not been stressed.

Most of us know that during precision creep testing, using 5-in. gauge length specimens, it is possible for the specimens to start necking in one place on the gauge length and then for the fracture to occur elsewhere. This is indicative of the statistical nature of creep failure and also such an effect must be capable of producing transitions of the type under discussion.

Apart from inviting Dr Glen's comments on this further mechanism for inducing transitions at a late stage in the creep curve, this particular phenomenon does introduce us to the subject of scatter. Much of the data which has been presented does not show evidence of scatter, and yet we are all aware that this is a major difficulty of creep testing. For example, we have recently done 16 tests on simple ferritic steel of the general type under discussion, and we found ductilities at rupture ranging from 16% to 35%. These are only short-time tests and it may be that in the long-time creep testing with which we have been mainly concerned in this discussion, that this type of scatter might be minimized. It is likely, however, that if sufficient testing had been carried out by Dr Glen to illustrate the importance of scatter, some of his conclusions might have been modified. In particular, this emphasizes Mr Kirkby's point that Fig. 12 in the paper on ductility might present a different aspect if sufficient tests to show the magnitude of scatter were included.

To determine what practical use we could make of this work, we examined some of our G.18B data from the point of view of transitions and Fig. *P* illustrates some of these results. We found that at the higher stress and hence shorter testing times, even a complex steel such as G.18B did not show any pronounced transitions. These transitions were confined to tests at very low creep rates. When the creep rates were higher the transitions were ironed out and relatively smooth curves resulted, such as the one for the test at 11tons/in² at 700° C as illustrated in Fig. *P*.

The shape of the two remaining tests illustrated in Fig. *P*, at 4tons/in² and 7tons/in² do show several of the transitions under discussion, and have a very complicated form. The point we should like to emphasize is that it is impossible for this steel under these conditions, to obtain any indication of the shape of the total plastic strain/creep rate curves at low stresses from the shape at high stresses. Extrapolation is therefore, in our opinion, impossible on this alloy at this test temperature.

In view of the substantial amount of data which Dr Glen has presented representing favourable cases to his argument, this must be regarded as something of an exception, but we would welcome Dr Glen's comments.

(The **Chairman** read a joint contribution by Messieurs **C. Crussard**, **A. Constant,** and **J. Plateau** (IRSID, France)): The excellent work of Dr Glen on the effect of precipitation on the high-temperature tensile properties of steel is well known in France, and we would congratulate the author on having extended his research to the more complex case of creep and at the same time proposing an extremely interesting experimental method.

As early as 1951, French investigators[7] had demonstrated that the creep test could be considered as an extremely sensitive method of physico-thermic analysis and recommended a study of the derivative curve, i.e. to use the curve of creep rate as a function of time to show the effect of the physico-chemical or structural changes produced during the creep test in the metal concerned. In studying the derivative curves as a function of plastic deformation instead of a function of time which perhaps increases still further the sensitivity, Dr Glen has shown here the great possibilities of the method as applied to the most difficult aspects of the study of creep, i.e. extrapolation, ductility at high temperature, and the influence of alloying elements.

What has particularly struck us in the collection of three important papers by the author is the lucidity with which he discusses the advantages and disadvantages of the proposed methods, their possibilities and their limits, and this physical examination which he makes answers many of the observations and remarks which could be put forward in a discussion.

There are, however, a number of points which are so important that we think they should be underlined. As far as the very difficult problem of extrapolation is concerned, we think that the proposed method represents definite progress compared to other methods at present in use, such as linear extrapolation on a double logarithmic diagram, the methods of Larson–Miller and of Dorn, because all these methods rely implicitly, whether it is stated or not, on hypotheses which are generally far too simplified to deal with the case of industrial alloys. Although these methods are not without interest, provided they are used intelligently, the method proposed

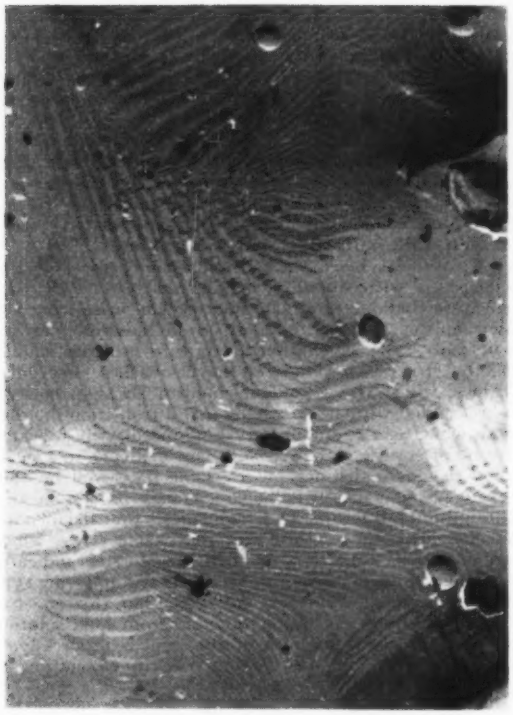


Fig. *Q*—Microfractograph of rupture surface of a specimen of Armco iron at — 196° C, first tempered; creep conditions 700° C and 1 kg mm² for 52 h
× 16 000

by Dr Glen, by taking account of the physico-chemical and structural changes in the test piece during the test, is certainly superior to them.

We would, however, point out that whatever the method used, an extrapolation always involves a fair degree of uncertainty which is not eliminated except by carrying out tests of extremely long duration. With the method recommended by Dr Glen, an extrapolation of the creep-rate curve is difficult because it is used either for stresses lower than that of the test or for lower temperatures, and it presupposes that the fundamental phenomena influencing the changes in the creep-rate curve are produced in the same way under these conditions. On the other hand, as has been pointed out by the author, it is quite certain that extrapolation is still more difficult for the case of rupture than for the case of creep because in this case it is not known at what elongation rupture will occur and the extrapolation must not exceed this elongation. Apart from the approximation made in extrapolating the strain rate curve, a further approximation must be made in fixing the value of elongation at rupture. It seems to us in this respect that a simultaneous study of curves of creep deformation against time to rupture, the strain rate curve, and the curve of elongation at rupture against time would lead to a more exact assessment of the approximations which should be made. However, the possibility of brittle fracture which is practically unpredictable always places some doubt on extrapolations, and we agree with the author when he concludes that long-time rupture tests at service temperature are necessary to be certain that a steel is in no danger of giving a brittle fracture.

As far as the mechanism leading to brittle fracture

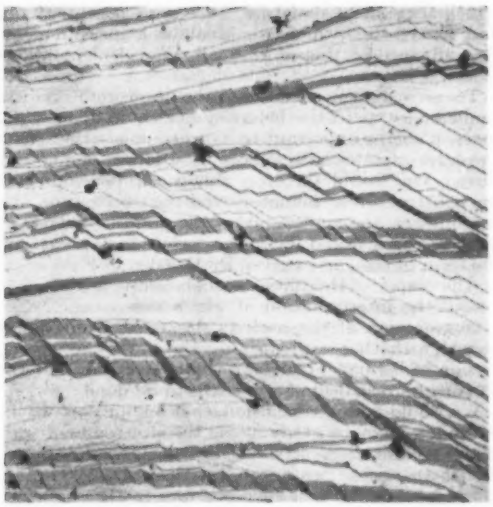


Fig. *R*—Microfractograph of intergranular rupture surface of a specimen of industrial nickel, previously weakened

is concerned, the theory presented by the author suggesting a delay in time between precipitation in the grain and at the grain boundaries seems to us extremely interesting. It is possible, however, that other phenomena may also be involved, in particular the phenomenon of intergranular absorption in the way which has been suggested by McLean and one of us. We think we have found some experimental confirmation by the micro-examination of fractured surfaces of test pieces for different steels previously subjected to creep tests. We give the example of a sample of Armco iron tested at 700° C for 52 h under a stress of 1 kg/mm². After creep the test-piece was broken in liquid nitrogen and we have observed with the electron microscope striations on the rupture surface (Fig. *Q*) very similar to those which we have observed previously on the fractures of a sample of industrial nickel which had been subjected to an embrittling heat treatment, and for which the embrittlement was related to the appearance of striations at the grain boundaries (Fig. *R*).

By analogy with the case of superficial absorption met in the study of deoxidation of metals,[8] we have put forward the hypothesis that these striations seen in the case of nickel resulted from absorption at the grain boundaries.[9] In the case of creep, the striations visible in Fig. *Q* result, we believe, from the phenomenon of absorption. However, it remains to be discovered whether the absorption occurs before or after the formation of a crack.

Finally, the application of the method proposed by the author to the study of the influence of alloying elements on creep appears to yield very interesting results. May we be permitted to make a suggestion which we believe will increase the interest of this investigation. As the author has said, the heat treatment and perhaps even pre-heat treatment given to the material has a considerable influence on the resistance to creep. Under these conditions, and to separate effectively the action of alloying elements and the effect of heat treatment alone, it would seem useful to examine carefully the structures involved because the same treatment could lead to different structures depending on the hardenability of the steel and the dimensions of the material treated. One could, for example, limit the study to the three

principal types of structure, i.e. ferrite–pearlite, bainite, and martensite, which would allow a measurement of the influence of this or that alloying element for each case.

To give a firmer base for the interpretation which could be given to the behaviour of these diverse structures, it seems important to us to complete the study of the form of creep-rate curves by a study of fine structures. One could therefore follow the development of the substructures, the nature and the distribution of the precipitates, the appearance of the grain boundaries etc. Such research would no doubt permit an explanation of the form of the curves given by Dr Glen in Figs. 5 and 6 of the paper "Ductility in high-temperature rupture tests," the interpretation of which seems complex.

Our opinion of the matters discussed is in complete accord with that expressed by the author at the conclusion of his papers, and we think with him that there remains a great deal of work still to be done.

We would add in conclusion, that on our part we have begun a systematic study, using the electron microscope, of the modifications of structure which occurred during creep.

WRITTEN CONTRIBUTIONS

Dr-Ing **A. Krisch** (Max-Planck Institut, Germany): Dr Glen has presented an interesting theory on the creep test which represents an important contribution to the well known works in this field. He relies in his theory on the experimental observations of transitions in the time/elongation curve.

Particularly important are his remarks on low-deformation rupture. He associates them with these transitions and indicates a method of finding, by extrapolation, the time/elongation curves for long-time stresses and simultaneously of determining data on deformability in the long-time test of about 100 000 h. If this method proves to be generally feasible it will mean a great step forward.

Dr Glen suggests that the development of transitions in the elongation/creep-rate curve is associated with changes in the structure, preferably with precipitations. It would appear, however, that a vast amount of meticulous detail work is still required in order to confirm the connections indicated by Dr Glen and to develop the theory along these lines. Our own tests over recent years confirm the theory to the extent that we established such structural changes in some steels during the creep test, but, unfortunately, no time/elongation curves of sufficient accuracy are available to establish the connection with Dr Glen's theory. In suitably heat-treated Cr–Ni–Mo steels, for instance, we were able to recognize the occurrence of molybdenum carbide Mo_2C after a testing time of 1 100 h at 500° C.[10] In Cr–Mo–V steels we found the ferro-molybdenum carbide Fe_3Mo_3C after the time rupture test of 14–240 h at 550° C while in the annealed condition only the vanadium carbide VC had been observed.[11] In austenitic Cr–Ni steels which were stabilized with titanium or niobium (columbium) it was possible to prove a sequence of reactions.[12] While before the creep test only the carbide of the stabilizing elements TiC or NbC was present, after an annealing time of some 1 000 h the carbide $M_{23}C_6$ appeared. The necessary carbon for this phenomenon was not previously recognizable in carbides, so was presumably dissolved in the austenite. After a further period of testing this carbide $M_{23}C_6$ disappeared again and the σ phase occurred and grew to substantial contents in prolonged time rupture tests. In addition to these principal reactions a certain part is played by the formation of secondary and tertiary carbides. The reactions described here were shown in our tests to be dependent on the temperature and on the load. There is reason to assume, therefore, in line with Dr Glen's theory, that with low loads and elongations the reactions progress more slowly and that, for instance, the dangerous formation of the σ phase remains within limits, i.e. the lower-loaded test specimen shows an adequate breaking time.

Dr **L. M. T. Hopkin** (National Physical Laboratory): I would like to congratulate Dr Glen on his observations concerning the changes of creep rate which have for long been known to occur during some creep tests. He has been able to show that these changes are related systematically with both stress and temperature.

My interest has been directed mainly to the paper concerned with ductility in rupture tests. Dr Glen suggests that intergranular fracture can arise from two mechanisms: in the first, which he described as the 'normal' type, elongation at rupture becomes smaller as the stress is lowered but never reaches very small values, whilst the second mechanism Dr Glen believes is associated with strain ageing and a transition in creep rate which can result in very low ductility at fracture. I am not sure that there is sufficient evidence to suggest that these are distinct processes; it may well be that strain ageing is merely enhancing a single process causing intergranular fracture. Dr Glen suggests that the increase in elongation at fracture with decreasing stress which occurs in long-time tests on the 0·5%Mo steel indicates a change of mode of fracture from the strain-ageing type to the 'normal' type. It could be that this effect is due to precipitation of molybdenum from solid solution so that the creep strength of the steel is progressively weakened with a corresponding rise in elongation at rupture.

On the other hand McLean[13] has definitely shown that two processes of intergranular fracture do exist in commercial high-temperature alloys. At high stresses he shows that narrow wedge-shaped cracks propagate along grain boundaries whilst at low stresses grain boundary cracks form by the linking together of isolated holes. He shows that the formation of wedge-shaped cracks is consistent with Stroh's[14] theory in which the criterion for a crack to form is that the elastic energy associated with the stress concentration produced by relative movement at an interface, such as grain-boundary slip, is greater than the sum of the elastic and surface energies of the smallest stable crack that could form. According to Stroh the stress concentration set up by a sliding interface is increased if dislocation sources within the grains cannot operate, e.g. if they are locked by strain ageing. I would, therefore, like to suggest that the fractures at relatively high stresses which Dr Glen has associated with transitions of creep rate and strain ageing may be produced by cracks of the wedge type. According to current ideas the cracks formed by the linking together of isolated holes are the result of precipitation of excess vacancies in the presence of stress concentrations. The decrease of creep rate associated with strain ageing in Dr Glen's results would lower the concentration of excess vacancies rather than increase it so it seems that this mechanism of fracture is unlikely to occur preferentially during a transition of creep rate.

There is another factor probably affecting ductility at fracture which is worth considering. Alloying additions which produce strain ageing have atoms with large size-factors compared with those of the basis metal and for this reason, also, the additions segregate to grain boundaries and lower the interfacial energy. This decrease of interfacial energy facilitates the formation of grain-boundary cracks of both types. Thus an alloy system exhibiting strain ageing would also be expected to rupture with low elongations in an intergranular manner. This I have shown to be the case in tensile

tests at elevated temperatures on single phase copper–antimony alloys[15] where strain ageing and severe intergranular fractures both occur. There is evidence to believe that these intergranular fractures are the result of precipitation of excess vacancies produced by plastic deformation.[16, 17] It may be of interest to note that in this alloy system ductility at rupture in tensile tests shows no tendency to increase at temperatures above that at which strain ageing ceases.

Finally, I am not clear from Dr Glen's suggestion how he would assess a material for service by tests at the service temperature. It is true that tests can be made at decreasing stresses until rupture no longer occurs during a transition of creep rate, but that is no guarantee that another strain-ageing process within the service life of a component will not enhance fracture with a low ductility.

Dr **K. W. Andrews** (The United Steel Companies Ltd): The four papers represent a notable step forward and I should like to support what was said at the meeting in praise of the authors. The paper describing the cantilever method leads one to suppose that this method would be particularly useful for high alloy steels and other special alloys. Dr Glen's work, on the other hand, is applicable to the interpretation of creep data in general, but raises some interesting questions as to the effects of alloy elements, and I should like to make some observations about these effects.

Figure 4 of Dr Glen's first paper and Figs. 5 and 6 of the third illustrate a point about the magnitude of the effect of an alloy element. Presumably the most effective alloy elements are those which, for example, move the curve DBYEF in Fig. 5 of the third paper, as far to the left as possible and a flat curve of this kind is preferable to a hump. The latter implies that the effect of the alloy element is more transient and the former that it is more sustained. Thus there are two parameters for each effect: one is the height and the other its width or duration in terms of strain (or time). The total effect is represented by an area, but this misses the separate significance of height and width. It would be interesting to know whether the effect of a given alloy element has been considered by the author in this way.

The author has discussed what structural changes or processes may go on whilst such transitions occur. It is clear, however, that it would be valuable to tie in his results with an extensive investigation (including electron microscopical study) to discover more precisely what precipitation processes are involved. Dr Glen's examples include tests taken at temperatures as low as 375° C on the one hand and as high as 600° C on the other. At the former temperature the time to reach the same metallurgical state would be very much longer than at the higher temperature. (A factor of say 10^8 might be involved.) Hence the peaks will be much sharper at the higher temperature even allowing for the fact that the stress applied may be considerably higher. My colleague Mr Murray pointed out that a number of metallurgical changes or processes could be involved and there is no reason why these should not be regarded as forming a sequence. Thus segregation of atoms to dislocations or other preferred sites could be followed by the formation of a coherent precipitate and then by fine particles of an incoherent precipitate. From then on, especially if the precipitate coarsens, the strength will decrease and the effect of the element has passed its peak. Hence the point *Z* in Fig. 6 (the third paper) could represent the stage at which the precipitation has become incoherent or exceeds a certain small particle size. (It is not necessary to assume that all of the strain-age hardening constituent is used up, since some of it can still remain in solid solution.)

It would thus appear that at lower temperatures a very considerable time elapses during which the segregation of atoms can play an important role, but that at higher temperatures such as 600° C the full precipitation stage is reached in a very short time. Hence if we consider the area under the peak at lower temperatures, it will extend over a longer time and presumably over a longer strain interval, and that part of it due to segregation alone would be much more significant. Some of these facts are supported by studies we have made in connection with the precipitation processes which occur in a number of alloy steels with or without the application of stress. The strength of a steel is maintained presumably because the progress towards a thoroughly over-aged condition is still slow at practicable service temperatures for a given steel, but also because there are sequences of precipitates which may or may not be independent of each other. These should correspond to successive peaks on Dr Glen's strain/rate curves. An interesting point is that if there is one process which has reached its peak and another starts there are two possibilities. The new precipitate can be separately nucleated and in this case there may be a pronounced new peak. Alternatively, if it is nucleated by the previous precipitate the extra hardening or retarding effect will probably be less and reveal itself by a change in direction in the falling off side of the peak (*ZE* in Fig. 6).

These observations are made in order to suggest one or two ways in which constitution and structure could possibly be linked with Dr Glen's interpretation of creep data. We hope that progress will follow along these lines and any others which promise to consolidate our knowledge in this field.

Mr **P. F. Tomalin** (The English Electric Co. Ltd): Dr Glen's method of analysing creep data is undoubtedly a great step forward in that it takes into account the effect of metallurgical changes occurring while the test is in progress. Those who have attempted to apply this new approach will have realized what a large amount of work must have been involved in the development of the theory, and Dr Glen is to be congratulated on the results he has achieved.

One aspect of the method of extrapolation appears to be open to doubt. In the prediction of creep curves Dr Glen assumes that the transitions in creep rate move along the total strain scale with changes of temperature and stress, the transitions occurring at lower strains as stress or temperature are reduced. It is also implied that each transition is due to formation of a particular precipitate incorporating one of the alloying elements present. However, it seems to be established that in some alloys a change of temperature may result in a new precipitate appearing, and at the same time one of the precipitates found at the original temperature may be suppressed. Under these circumstances, transitions in the strain/rate curve may disappear with change in temperature rather than be shifted to lower or higher strains. If this happens, the extrapolation by drawing a set of similar 'profiles' over a range of temperatures will be incorrect. When the behaviour of the precipitates concerned is more completely understood it may be possible to apply the necessary correction to the shape of the strain/rate curve for specific temperatures.

In connection with rupture ductility, I wish to refer to a test result which is of interest. A creep test was carried out at 565° C on a specimen transverse to a Mo–V steel weld deposit, i.e. the gauge length contained a length of weld metal in the middle and parent metal at each end. The test conditions were such as to give a very low creep rate, i.e. the total creep in 100 000 h would not have been expected to exceed 0·1%. After the test had been in progress for 25 000 h it was decided

that it should be stopped. To simulate the effect of overloading in service the stress was increased to a value which gave a fast and increasing creep rate. From the time the stress was increased, failure occurred in 30 h. The specimen ruptured in the centre of the Mo–V weld metal, with an elongation on 0·56 in. of 80%, and if the same extension is based on a 2-in. gauge length the elongation is 23%.

It is normally assumed that the rupture ductility of Mo–V steel over long periods of time at 565° C is less than 10%, if the test is run to failure at constant stress. The result of the test described, although unconfirmed, does indicate that if the strain rate is low then the reduction in ductility, insofar as it affects service performance, may be far less severe.

I would welcome Dr Glen's opinion as to whether, in the same way as he advocates rupture testing at the service temperature, he would agree that the tests should be carried out at the service strain rate, and then ruptured at specific times by increase in the stress.

Dr **S. A. Main** (Hadfields Ltd): The success of Dr Glen's efforts will, it is to be hoped, stimulate similar efforts by others towards the much-needed rationalization of the problem of the behaviour of steels in creep. In particular, he has shown that rate of creep can with advantage be related to total creep rather than in the usual way to time, which marks a distinct step forward through the useful indications and conclusions which, as he has shown, can so be derived.

I have been able to pursue my own investigations in this direction only as occasion served between other work. I hope to publish a description shortly, and believe it will be found to support, and help to extend, Dr Glen's valuable work. Some of the results and conclusions are as follows.

In his first paper Dr Glen points out that, in an alloy, and in the initial or primary stage, the variations in rate of creep resemble those of a pure metal. For a limited number of highly alloyed steels of the heat-resisting class this fact had been reinforced[18] to the extent of showing that the formula established for pure metals applies equally well to those steels.

Subsequent application to other compositions of heat-resisting steels has confirmed this, indicating the probability of its more general application. If this is so, much of the desired rationalization of our creep testing has already been effected, at any rate in the initial stages. It would be interesting to know whether the author has himself followed this line of investigation, and whether he can confirm this from his own observation.

Useful results accrue, e.g. in settling the position of the point B (Figs. 2 and 4) as the stage at which the rate of creep abandons the Andrade formula; point B being the effective beginning, which it is very desirable to know, of the transformation in the steel responsible for this departure.

A trial application of Dr Glen's method of plotting to a series of increasing stresses applied to the same heat-resisting steel and at one temperature had unfortunately not the same success, i.e. the curves do not form a logical sequence, rather confirming the author's experience (Fig. 29 of the third paper) in one case. Doubtless the explanation then afforded, i.e. the difficulty of reproducing by heat treatment, in highly complex alloys, a particular structure, is the correct one. The lesson to be learned, or rather to be emphasized since metallurgists are already aware of it, is that stricter control of heat treatment, as regards both time and temperature, is very necessary if complete regularity of performance in service at high temperature is to be ensured.

Dr Glen comments on p. 30 of his second paper on the drawback to progress in the lack of a theory to explain those low elongations which can be obtained with complex alloys in certain circumstances. It is true, of course, that no such explanation is in current circulation, though I would call attention to one which has been available for many years,[19] and which has so far not received the attention which I think it deserves. Based as it is on solid ground in the well known austenite–martensite–ferrite carbide series of transformations, it attributes the tendency to brittle fracture to a temporary preponderance of the martensitic constituent. This applies both to brittle tendency in creep and to temper brittleness. I believe that only by this means can clarity at last enter into the understanding, not only of low elongations but also of the more general connection between the processes of recrystallization and behaviour under stress.

In studying by microexamination the fractures obtained in rupture tests described in the second paper, the author rather appears to imply that a certain 'cup-and-cone' fracture obtained in creep was essentially different in its mechanism of formation from any of similar appearance in an ordinary tensile test. His reason is that intercrystalline cracking has occurred at the centre in the former case; but, as has been shown,[20] this is the case also in the ordinary tensile test.

As regards the smoothing of curves, I personally find of advantage any method of plotting which brings the curve or any desired portion of it more nearly straight. Smoothed values of the creep strain thus more readily obtained are easily replotted linearly for determining the tangent slope. In the case of the double logarithmic plot the rate of creep is obtained without further plotting by using $\mathrm{d}C/\mathrm{d}t = C/t \times \mathrm{d}\log C/\mathrm{d}\log t$.

Mr **T. J. Heal** (UKAEA): I should first like to pay tribute to Dr Glen's distinguished work in the field of creep. A wealth of accurate data has come from his laboratory over a period of many years and his published papers always show his desire to probe into the underlying causes of the behaviour he has been observing. He has made use of both analysis and imagination in his search for underlying themes to allow worthwhile predictions of behaviour in areas where experimental results are not available. The three papers which have been presented are perhaps the best examples of his approach and they have made very valuable reading.

There are a number of points where I should like to comment. Dealing firstly with his paper "A new approach to the problem of creep," and the concept which he introduces of plotting the log of the total plastic strain against the log of the creep rate, it is not clear to me why the changes in the metastable equilibrium responsible for the marked changes in creep rate should be associated with particular levels of total plastic strain.

We had some discussions with Dr Glen 18 months ago and we have attempted to apply his methods to some of our own results. We have found some difficulty in deciding what are significant changes in creep rate and have tended to get differing results in some cases when the plots have been carried out independently by two or three different members of the research staff. I do not wish this comment to be taken to imply that his approach falls down in practice, I merely wish to make it clear that in some cases it is difficult to apply. We consider the effort involved well worthwhile because we entirely agree with the criticisms which Dr Glen has made of the more straightforward extrapolations and it is clear to us that any approach is of value if it provides the possibility of detecting time-dependent phenomena responsible for significant changes in creep resistance. We feel so strongly about this that we have mounted a considerable research programme designed to detect time-dependent changes in possible pressure-vessel

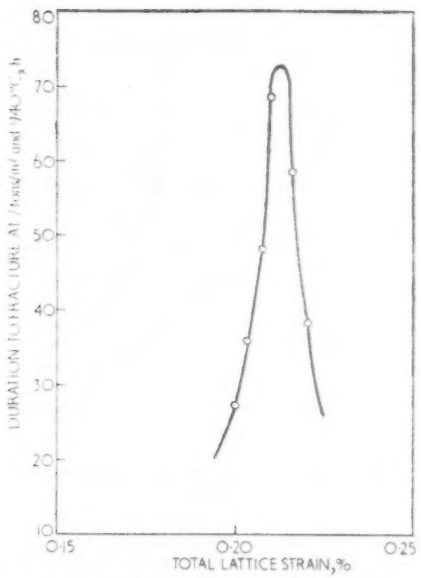


Fig. *S*—**Relationship between total lattice strain and creep rupture strength in Fe–Co–Cr–Ni alloy hardened with Ti, W, Mo, and B**

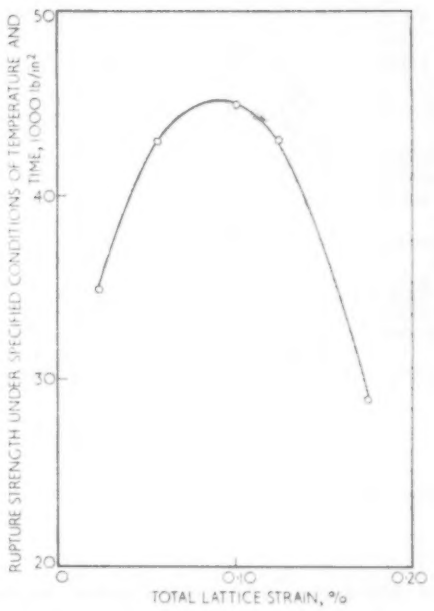


Fig. *T*—**Variation of creep rupture strength of V-bearing ferritic steels with total lattice strain values, calculations based on data given in ref. 23**

steels. We are investigating the effect of temperature and of temperature plus strain over a range of values of these parameters and are examining the specimens for changes in hardness, tensile properties, and microstructure, the latter by both optical and electron microscopy.

Turning to Dr Glen's paper on the effect of alloying elements, I should like first to confirm the claim he makes in relation to Fig. 14, that his prediction of behaviour at 375° C was made before he was supplied with our data at 380° C, which is shown in the same figure, and where, as Dr Glen points out, there is reasonable agreement.

The aspect of this paper about which I have some reservation is the interpretation which Dr Glen attaches to the changes in slope in his families of strain/rate curves. We are unhappy at giving unqualified acceptance to his interpretation without more evidence. Dr P. E. Brookes of the Culcheth Laboratories has gathered some evidence that solid solution effects can be of significant importance in determining creep resistance first with alloys based on materials other than iron, and second with steels. Figure *S* shows the relationship between total lattice strain and creep rupture strength in an iron–cobalt–chromium–nickel alloy hardened with titanium, tungsten, molybdenum, and boron. Figure *T* shows the same effect for vanadium-bearing ferritic steels. In each case it can be seen that the maximum resistance to creep rupture is developed for less than the maximum total lattice strain.

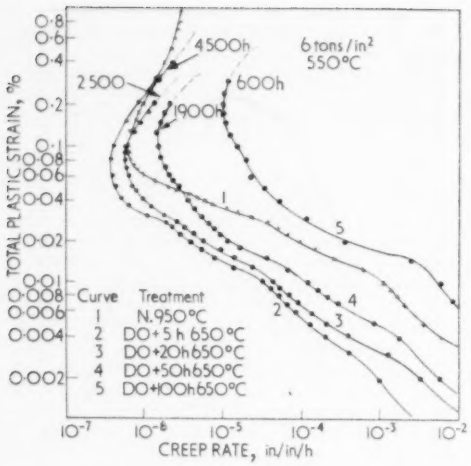
We believe that these results are associated with clustering effects in non-homogeneous solid solutions. If we examine Dr Glen's suggestion that the inflections are due to the precipitation of carbides, nitrides, etc., then in the case of complex austenitic alloys which contained molybdenum and/or tungsten, together with titanium and/or niobium, once the titanium–niobium carbide has formed, the difference in heats of formation would appear to prohibit subsequent formation of the tungsten molybdenum carbide.

Turning to Dr Glen's third paper on ductility in high-temperature rupture tests in the hypothesis which he advances on pp. 35 and 36, I cannot follow his argument that the diffusion of alloying element such as molybdenum to a dislocation occurs more readily in the grain than at the grain boundary. In an homogeneous solution there would appear to be equal availability of molybdenum atoms in all points. A point of rather greater importance is the inference that brittleness, which takes a long time to develop, is not so deleterious as is generally supposed because of the possibility of stress relaxation before the brittle state arising. There are a number of cases in our experience where stress concentrations do not relax in service, and this argument does not apply. The sort of case I have in mind is that where, due to the geometry, a permanent stress concentration is set up; as an example, one can cite the case of entry pipes in a pressure vessel where the surface of the cylinder is interrupted.

AUTHORS' REPLIES

Dr **Glen**: I am extremely grateful for the very many kind remarks made about these papers and I thank the various contributors most sincerely. I was pleased to note, however, that a few of my friendly rivals did not attempt to pull their punches. This serves to emphasize the many difficult problems which still confront us and have to be solved before we really know what happens during creep testing.

Dr Benson very rightly quotes some of the pioneering efforts by R. W. Bailey and shows how correct he was in many of his ideas and how these ideas lead up to the present work. The strain/rate curves which Dr Benson shows for a zirconium–copper alloy with and without molybdenum are extremely interesting and serve as a valuable confirmation of the conclusions in the paper as applying to non-ferrous alloys. The addition of



Curve	Treatment
1	Normalized at 950° C
2	" " " + 5 h at 650° C
3	" " " + 20 h " "
4	" " " + 50 h " "
5	" " " + 100 h " "

Fig. *U*—Effect of tempering at 650° C on 0·5%Mo steel 18

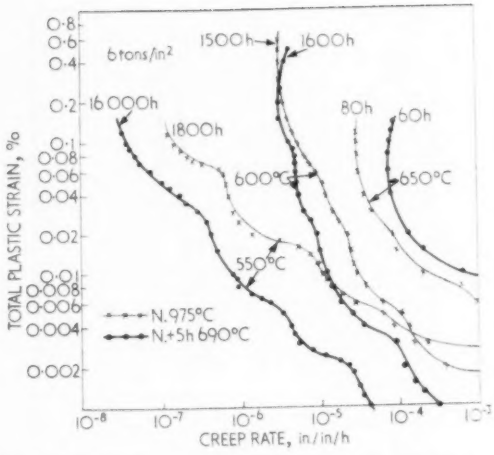


Fig. *V*—Effect of tempering on creep behaviour of Mo–V steel

molybdenum results in solution hardening and also possibly in the presence of a coherent precipitate before testing so that the initial creep resistance of the molybdenum-containing alloy is increased. Thus, since the creep rate decreases more rapidly in this alloy, more dislocations are stationary at a low strain level with the consequence that transitions in creep rate can occur at a lower strain level than in the corresponding alloy without molybdenum. A similar though less pronounced effect can be obtained on tempering 0·5%Mo steel, as is shown by the strain/rate curves in Fig. *U*.

In reply to Mr Murray, the choice of the terms 'strain-age-hardening' and 'strain-induced-precipitation' was quite deliberate since these effects occur only during tensile testing at high temperature or during actual creep testing. Mr Murray, however, suggests that transitions result from already well known phenomena and reduces the possibilities to two, namely, 'atmospheres' and 'coherency strains associated with precipitation.' With regard to atmospheres, I have already admitted, as shown in the final conclusion in the third paper, that this phenomenon might explain some of the transitions. In the meantime, however, I consider it preferable to use only one explanation in the interests of simplicity, if nothing else. Mr Murray then states that the creep strength of 0·5%Mo steel increases up to an optimum value when coherency strains are at a maximum and that beyond the optimum the Mo_2C precipitate loses coherency and begins to overage. Figure *U* shows a set of strain/rate curves indicating this phenomenon but it should be realized that the maximum coherency strains obtained after about 10 h tempering at 650° C increase the initial creep resistance to a maximum. What we are concerned with, however, is the latent creep resistance, i.e. the effect of molybdenum in causing a transition in creep rate. If we re-examine the evidence given in the paper the difference will be made quite clear. This evidence refers to tests carried out by the National Physical Laboratory and not by the author. In Figure 17 of the second paper, it will be seen that in a creep test at 650° C and a stress of 1 ton/in², the transition due to the presence of molybdenum begins after several hundred hours of testing. This transition is quite marked and reduces the creep rate by a factor of about 5 despite the fact that the time at 650° C is very much greater than the 10 h tempering at the same temperature to give a maximum in the initial creep resistance. In Fig. 6 of the first paper the same steel was tested at a temperature of only 500° C and a stress of 6 tons/in². After about 40 000 h testing a pronounced transition began. Since 40 000 h at 500° C is equivalent to about 40 h tempering at 650° C it is clear that, even in this case, the transition occurred at a time in excess of that required to give maximum coherency strain. Going to the other extreme the results of tensile tests[21] show that molybdenum has a maximum effect in a test lasting only 1 h at 500° C. I think, therefore, that Mr Murray must admit that transitions occur at times which bear no direct relationship to the times for maximum coherency strain during tempering. The effect of strain, time, and temperature must all be considered so that the term 'strain-age-hardening' or 'strain-induced-precipitation' seem appropriate until we know more about this complex phenomenon. I must add, however, that some of the initial creep resistance can be converted to latent creep resistance by a suitable heat treatment and vice versa. Thus we are not dealing with an entirely new phenomenon. It simply means that the phenomena Mr Murray mentions are not quite as well known as he suggested.

With regard to Fig. 5 of the first paper I fully realize that 2 h tempering at 650° C will partly spheroidize the Fe_3C and will cause diffusion of manganese into the carbide. The fact remains, however, that three transitions are obtained even in a test at 450° C, and these transitions appear to be associated with the presence of nitrogen, carbon, and manganese. However, as will be realized from the arguments given above with reference to molybdenum, much further work will be necessary before we know the actual type of dislocation precipitates which form in any particular case. Similarly with regard to Mo–V steel the precipitates which form during testing must be similar to those in a carbon–manganese steel with the addition of a transition due to molybdenum. Tempering of Mo–V steel alters the initial creep resistance in a similar manner to 0·5%Mo steel. This is illustrated in Fig. *V* showing tests at 550°, 600°, and 650° C. With regard to Mo–V–Ti steel I agree that the bulk of the titanium carbide and nitride is out of solution at 1 150° C

but sufficient goes into solution at this temperature to cause transitions in creep rate. By heat treating at a still higher temperature a quite fantastic creep strength can be obtained for such a simple alloy, although unfortunately the material becomes brittle. This is surely due to further solution of titanium carbide and nitride. I note that Mr Kirkby agrees with me on this question.

With regard to intercrystalline failure Mr Murray takes me to task regarding the assumption that in complex alloys the same creep laws apply to grain-boundary creep as to the total creep. There is not much evidence to prove this, but McLean shows that this assumption is correct for a copper–beryllium alloy. This work is referred to in Fig. 11 of the second paper.

Figures 7 and 8 in the second paper certainly do not show actual transitions due to molybdenum, but the evidence given for molybdenum at other stresses and temperatures does show that transitions do occur and that the strain level shown is about correct. For the reasons already given I cannot agree that the minimum ductility occurs when coherency strains are at their maximum. Nevertheless I quite agree that the explanation of this whole phenomenon is still very sketchy and that the true explanation may be quite far removed from that suggested.

Mr Murray also does not agree with the suggestion that chromium delays the molybdenum transition and quotes the evidence of tempering. It has already been shown that such evidence bears little or no relationship to the latent creep resistance. Experimental evidence does seem to suggest that chromium delays the molybdenum transition. Finally, Mr Murray suggests that the stress at which the minimum ductilities are obtained approximates closely to the 0·1% proof stress at the temperature of testing. If this is so I would consider it a coincidence. In any case the whole purpose of the papers has been to elaborate on the effect of warm working and break it up into its component phenomena.

In reply to Dr Heslop I would agree that in Nimonic alloys the metallurgical evidence may show nothing to support conventional strain ageing. However, I think that the evidence of creep tests is much more sensitive and does show that at least three transitions can occur in such alloys. I quite agree with Dr Heslop that by altering the heat treatment the elongation at fracture of any alloys can be varied markedly. I think it will be found, however, that in such cases the whole shape of the strain/rate curves will have been altered so that the possibility of embrittlement must be decided in each case.

Dr. Bandel asks about negative creep. In the simple alloys which I have tested I have found no evidence of such phenomena, although there is no doubt it can occur in the more complex alloys. Where it occurs the strain/rate method of plotting cannot be applied unless a correction is made and much work would be required to show what the correct factor should be in any particular case.

Mr Smith would like to know the usefulness of short-time tests and acceptance criteria. As a general rule I would suggest that short-time tests are of little use since they may not differentiate between good and bad material. In certain cases, however, a short-time test can be developed which is useful, for example the NPL tests at 8 tons/in^2 and 450° C for carbon steel. Reliable extrapolation depends on many factors and I would suggest that where possible the tests should be of the order of one-fifth of the total extrapolated time. In some cases, of course, much longer time tests might be necessary depending on the complexity of the material.

Dr Baker has expanded on the possible effect of atmospheres and points out that atmospheres can be formed even with atoms which do not go into interstitial solution. If his views are correct it would imply that transitions can occur in simple solid-solution alloys and in my reading of the literature I never found any such effect. As I have already stated in my reply to Mr Murray latent creep resistance is not easily explained by the presently accepted theories of atmospheres or coherent strains.

I am afraid I must disappoint Mr Mercer regarding the prediction of the amount of precipitation necessary to prevent failure under given conditions. The effect of combination of alloying elements is so complex that it is not yet possible to produce a finished steel without doing some preliminary testing.

Mr Draper asked about the magnitude of the effect of creep on the strain-age-hardening factor, i.e. latent creep resistance. This will vary enormously depending on the alloy. In some cases it reduces the creep rate only a small amount, but in others, it can reduce the creep rate by as much as 10 000 times. Much work will be required before we know the exact relationship between latent creep resistance and ordinary precipitation hardening. Mr Draper found no transitions in a test carried out on Mo–V steel and suggests that if they take place they must take place below about 0·02% strain. This is indeed the case as is shown in Fig. 1', and is to be expected in view of the fact that his steel was tempered. With regard to the 2¼%Cr–1%Mo steel, the very fact that initially it creeps faster, i.e. it has a low initial creep resistance, ensures that the transitions will occur at high strain levels. The strain level at which the so-called tertiary creep seems to occur is, of course, no criterion as to whether or not a steel is good or bad. A family of curves in each case is necessary before it can be shown which of two materials is liable to be the better for any particular type of service.

I thank Mr Arrowsmith for presenting evidence of precipitation into the dislocations during high temperature straining.

Mr Kirkby mentions that some transitions are not very clearly defined and this, of course, is quite correct. Some ' wiggles ' as he calls them are no doubt caused by experimental error and it is only by studying the results of a number of tests on the same material that one can be quite certain that a small transition has occurred. If a transition is not very clear it can be neglected, although it means that the strain/rate curve will give an underestimate of the properties of the material since the transition becomes more pronounced the lower the temperature of testing. The absolute accuracy of the proposed method of extrapolation is, of course, difficult to assess without doing a tremendous amount of long-time testing. There seems no doubt, however, that it has many advantages over the older methods and no doubt as time goes on it can be made much more precise.

Figure 12 of the second paper is not based on one set of results, but on an average of about ten casts in each case, except that of the carbon steel. There is no doubt that Cr–Mo steel can, in some cases, behave similarly to C–Mo steel. For OH steel I have only found this true by overheating the material during normalizing. If Mr Kirkby obtains a straight line with stainless iron containing molybdenum, this would indicate some form of recovery or recrystallization, which as the paper mentions, would lead to higher elongations. In such cases I would expect the normal behaviour to occur at lower temperatures. Mr Kirkby, however, should realize that this paper on ductility is only a first step and has still very much further to go before we can control this rather important phenomenon.

The question of scatter is raised by Mr Kent. This, of course, was not raised in the paper to any extent since it is a very big subject and would require a paper in

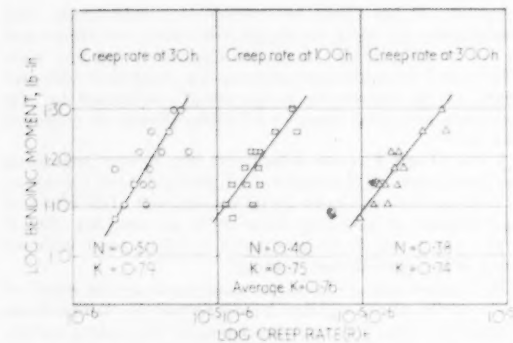
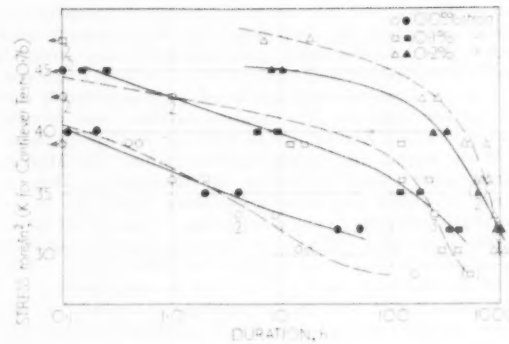


Fig. *W*—Log (M) log (R) plots at various times for H53-type alloy at 500° C



Solid points and solid lines indicate tensile tests.

Fig. *X*—Comparison of tensile and cantilever creep tests on H53-type alloy at 500° C

itself. I have no doubt that most of the scatter is due to metallurgical variables and the strain/rate method of plotting should pinpoint whether it is due to a variation in the initial creep resistance or due to a variation in latent creep resistance. Only when we know what causes scatter can steps be taken to minimize it. I agree that the evidence on carbon steels shows that at a given temperature in 100 000 h the plastic strain may vary from 0·03% to as much as 1·0%. However, by taking a safety factor on temperature of quite small magnitude the chances of obtaining a high deformation are quite remote.

Mr Child mentions another phenomenon which can cause a false transition, multiple necking. This, of course, is seldom if ever obtained with the simpler alloys and is not an argument that all transitions are false. In any case I do not think it would make a very pronounced transition and its presence would be detected by the shape of the test-piece so that a smooth curve could be drawn ignoring this false effect. I quite agree with Mr Child that if many more tests had been carried out showing the effect of scatter some of the conclusions in the paper might have to be modified. I might add, however, that I tried to avoid any conclusions which might be affected by scatter, and in many cases, the strain/rate curve reproduced was only one of many available. Figure *P* presented by Mr Child illustrates the difficulties of testing very complex materials which require an enormous time to show up the presence of transitions. However, I think Mr Child could extrapolate his data quite easily in the knowledge that it would be an underestimate of the properties. Even an underestimate is better than no estimate at all. I would not consider Mr Child's example an exception. It is simply that his tests were not carried out for a long enough time.

Messieurs Crussard, Constant, and Plateau make very many excellent suggestions regarding future work and I thank them most sincerely.

Dr Krisch quite rightly concludes that a great deal of further work will be necessary before the changes in structure resulting in transitions are clearly understood. I was very interested in his examples showing that a different carbide precipitated under creep conditions compared with the same material given only a long tempering treatment.

Dr Hopkin suggests that intergranular fracture might result from one mechanism. I agree that the actual mode of fracture may be the same. It is the effect of strain-ageing which results in a different behaviour. In the method proposed for assessing material for service, it is presupposed that the number of transitions is known so that the possibility suggested by Dr Hopkin should not occur.

Dr Andrews rightly points out that the most effective alloying elements are those which move the strain/rate curve as far to the left as possible, although both height and width must be taken into consideration. The effect of alloying elements is so complex and dependent on the other alloying elements present that it would be difficult to show precisely the effect of one particular element. I have not yet attempted such a study in any detail. With regard to structural processes, Dr Andrews elaborates on those suggested by Mr Murray. While what he says is all very correct, there still remains the difficulty of explaining the difference between initial and latent creep resistance as was mentioned in my reply to Mr Murray. Many of the suggestions made by Dr Andrews will be very helpful for future work.

It is suggested by Mr Tomalin that in some alloys a change in temperature may result in a new precipitate appearing. Such alloys will certainly require special treatment but the phenomenon would be clearly shown by the shape of the strain/rate curves obtained. The result of a rupture test on a Mo–V steel weld deposit mentioned by Mr Tomalin is extremely interesting and seems to support the conclusion in the paper regarding failure in service. I certainly agree that some tests of this type would give us much valuable information about rupture ductility.

I was extremely interested in the remarks made by Dr Main and look forward to seeing the full details in print. I heartily agree with him that strict control of heat treatment is very essential if uniformity of performance is to be obtained. With regard to smooth curves, I am afraid that although I attempt to smooth out experimental error as much as possible I very seldom find a creep curve which has a straight portion. This of course is evident from the shape of the strain/rate curves obtained.

Mr Heal would like to know why the metastable equilibrium resulting in transitions is associated with particular levels of total plastic strain. I prefer to think of the transitions as occurring at particular levels of creep rate depending on temperature, though the level of creep rate also changes with the stress imposed. A full explanation has, of course, still to be obtained. I quite appreciate the difficulties of deriving strain/rate curves, particularly at low stresses and temperatures. In the appendix to the first paper I have tried to suggest

TABLE 1

Comparison of tensile and cantilever creep strengths for H53-type alloy at 500° C

Tensile stress, tons in^{-2}	Cantilever creep stress for same strain time criterion, tons in^{-2}			Error of cantilever stress, %		
	0·05%	0·1%	0·2%	0·05%	0·1%	0·2%
32	29·4	30	31·0	− 8	− 6½	− 3
35	34·7	36·3	37·6	− 1·0	+ 3½	+ 7½
40	40·3	41·3	42·3	+ 1·0	+ 3½	+ 6
45	...	44·0	47·8	...	− 2	+ 6

the best method of carrying out this work. I would emphasize, however, that without a set of graduated curves it is very difficult to judge the data using the size of graph paper required. I am sorry if I gave a wrong impression regarding the effect of solid-solution effects. These no doubt may be quite considerable in some cases, but the main point is that they should not cause transitions in creep rate. In all the alloys tested, solid-solution effects are quite important although impossible to separate from the effect of ageing. With regard to the paper on ductility, more dislocations of a kind are present at the boundaries and thus presumably more molybdenum will be necessary in the boundary regions. It was for this reason that I stated that diffusion of alloying elements to a dislocation occurs more readily in the grains.

In conclusion, I would like to emphasize once again the very large difference between initial creep resistance and latent creep resistance. Many of the contributors attempted to draw conclusions between the effect of tempering and the behaviour during creep. Tempering, however, only influences the initial creep resistance and I would suggest that a more intense study of the phenomenon of latent creep resistance is now required.

Mr **Child** wrote: A further example of a detailed comparison between precision tensile creep testing and the cantilever bending method is given in this note. The data relates to a 11%Cr steel containing 5% of cobalt and other alloying additions.

A series of cantilever creep tests at various bending moments was carried out at 500° C and the log creep-rate/log bending moment plots at various stages during the tests are recorded in Fig. *W*. The slope of these curves decreases slightly with time, and so the creep exponent n falls from 0·5 at 30 h to 0·38 at 300 h while the cantilever creep parameter K similarly decreases from 0·79 to 0·74 over this time period. The average value of K is 0·76.

Using this value of K, the results of these cantilever creep tests are compared with precision tensile tests in Fig. *X* and Table 1.

There is a close agreement between the two testing methods and in no case do the results differ by more than 7½% when compared on the basis of stress to cause the same strain/time criterion.

These results may be contrasted with those given in Figs. 7 and 8 and Table III of the paper,[22] where a somewhat more complicated treatment was necessary.

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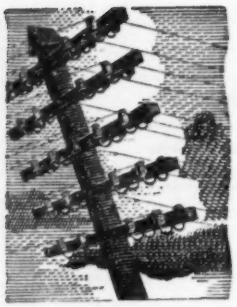

N E W S

Announcements and News of Science and Industry

THE IRON AND STEEL INSTITUTE

Special Meeting on the North-East Coast

The Institute is holding a Special Meeting on the North-East Coast from Monday to Friday, **6th to 10th July**, by invitation of companies in the district.

Visits will be made to the works of the following companies: Ashmore, Benson, Pease and Co., Consett Iron Co. Ltd., Dorman Long (Steel) Ltd., Head Wrightson and Co. Ltd., South Durham Steel and Iron Co. Ltd., and Skinningrove Iron Co. Ltd. In addition there will be a number of social functions. An attractive programme of excursions has been arranged for the ladies.

Full details of the programme were given in the June issue of the *Journal* (pp. 171–172).

"Stal in English"

By arrangement with the Department of Scientific and Industrial Research, The Iron and Steel Institute is undertaking the monthly publication of a complete English-language version of the Russian steelmaking journal "Stal". The January, 1959, issue is now available from the Institute, and future issues will be appearing at regular intervals.

The rates of subscription are as follows:

Published price

12 monthly issues £20 (plus 12*s.* postage)
Single copies £2 (plus 1*s.* postage)

Special rate for Members of The Iron and Steel Institute

12 monthly issues £15 (plus 12*s.* postage)
Single copies £1 10*s.* 0*d.* (plus 1*s.* postage)

Single copies are on sale as specimens if required.

6th Report of the Corrosion Committee

The Sixth Report of the Corrosion Committee has recently been published as No. 66 in the Institute's Special Report Series. Written by Dr J. C. Hudson on behalf of the Committee, the new report contains a full survey of the activities of the Committee since the appearance of its Fifth Report twenty years ago.

The Report covers the corrosion of iron and steel in air, soil, and water, details the results of numerous experiments, and recommends methods of protecting ferrous metals against atmospheric corrosion, soil corrosion, marine corrosion, and corrosion by industrial waters.

Copies of the Report, which is bound in cloth boards and illustrated with 39 drawings and photographs, may be obtained from the Secretary, price £3 3*s.* 0*d.* to both Members and non-members.

NEWS OF MEMBERS

Mr **W. E. Bardgett** is President-Elect of The Institution of Metallurgists.

Dr **Howard Biers** has been named senior consultant to Union Carbide International Company.

Mr **D. Campbell** has left Brymbo Steel Works Ltd. to take up an appointment with Richard Thomas and Baldwins Ltd., Redbourn Works, Scunthorpe.

250th ANNIVERSARY OF THE SUCCESSFUL USE OF COKE IN STEELMAKING

Anniversary Meeting in Great Britain, 23rd to 25th September

In 1709 Abraham Darby achieved the first successful use of coke in ironmaking at Coalbrookdale. To celebrate the 250th anniversary of this event, a meeting will be held from Wednesday to Friday, **23rd to 25th September, 1959,** at the University of Birmingham and at Coalbrookdale, Shropshire.

Allied Ironfounders Ltd. have made arrangements to preserve the furnace and to establish a museum on the site of Abraham Darby's experiments at Coalbrookdale. A committee has been set up under the chairmanship of Sir Frederick Scopes to organize the meeting.

The following is an outline programme of the meeting:

Wednesday, 23rd September

Evening

Opening lecture at the University of Birmingham by Professor T. S. Ashton on "*The economic and social background of the iron industry in the eighteenth century*"

Thursday, 24th September

All day

Presentation and discussion at the University of Birmingham of the following papers:

"*The charcoal iron industry in Great Britain, with special reference to Shropshire and the development of the industry up to* 1750," by B. L. C. Johnson

"*The contributions of the Darby family to ironmaking and the iron industry,*" by Dr A. Raistrick

"*Coalbrookdale* 1709," by Dr R. A. Mott

"*Ironfounding past and present,*" by Dr J. G. Pearce, C.B.E.

Evening

Dinner at the Grand Hotel, Birmingham.

Dress: Dinner jackets

Friday, 25th September

All day

Visit by motor coach to the Coalbrookdale furnace and museum, the Iron Bridge, and other historical sites in the neighbourhood.

Further details and forms of registration may be obtained from the Secretary, The Iron and Steel Institute.

Mr **L. S. Carstairs** has been appointed a director of the Barrow Hematite Steel Co. Ltd.

Dr **B. Cina** has left the Brown-Firth Research Laboratories to join the Israel Atomic Energy Commission (Metallurgical Laboratory), Hakiriya, Tel Aviv.

Mr **H. H. Egginton** has been appointed general manager of British Ropes Ltd., Doncaster Wire Mill.

Dr **B. S. Ekelund** has been appointed senior research supervisor of ceramics and abrasives in the chemistry research department of the Cincinnatti Milling Company.

The **Earl of Halsbury** has been appointed to the board of Lancashire Dynamo Electronic Products Ltd., as chairman.

Mr **J. E. Holgate** has accepted a six-months assignment in Argentina at the request of the United Nations Technical Assistance Administration and will make his headquarters in Buenos Aires. He will advise the Argentine Government on matters concerning the development and utilization of iron-ore deposits and iron production.

Mr **P. Heptonstall** is now assistant sales manager of the Park Gate Iron and Steel Co. Ltd.

Mr **C. R. W. Hughes** has joined British Ropes Ltd., as group general export manager.

Mr **C. W. Kieft** has been appointed a director of the Barrow Hematite Steel Co. Ltd.

Mr **C. McNair** is now a local director of Millspaugh Ltd.

Mr **W. R. Owen** is now responsible for the Midlands operations of Honeywell Controls Ltd. and has moved to Birmingham.

Mr **A. J. Peech** has been appointed deputy chairman of the United Steel Companies Ltd.; he will continue to act as general managing director.

Mr **W. J. H. Rutherford** is now works director of the Clyde Engineering Co. Pty. Ltd., Clyde, NSW, Australia.

Mr **J. D. Sharp** has moved from the research and development department of the United Steel Companies Ltd. to Steel Peech and Tozer.

Mr **R. H. Siddons** has been elected secretary of the Lincolnshire Iron and Steel Institute.

Dr **W. Steven**, superintendent of the development and research department laboratory of The Mond Nickel Co. Ltd., has transferred to the development and research division of The International Nickel Company Inc., New York, as director of research. He has also been elected an assistant vice-president of the company.

Mr **J. T. Watson** has been appointed manager of the light section mill at the Dalzell Steel Works of Colvilles Ltd.

Mr **J. G. Wistreich** has been awarded the degree of Doctor of Science (Engineering) by London University.

Obituary

Mr **William Beck Baxter**, C.B.E. (*elected* 1933; *Member of Council* 1945; *Vice-President* 1947; *Honorary Vice-President* 1952), of Scunthorpe, on 8th May, 1959.

Sir **Archibald John Boyd**, M.A. (*elected* 1914), of London, on 11th May, 1959.

Mr **Murdo Mackenzie**, M.B.E. (*elected* 1945), of London, on 5th February, 1959.

Mr **Thomas Middleton** (*elected* 1905), of Sheffield, at the age of 94.

Monsieur **François Perot** (*elected* 1938; *Honorary Vice-President* 1958), of Ougrée, Belgium, on 16th May, 1959.

AFFILIATED LOCAL SOCIETIES

North Wales Metallurgical Society

The following officers and committee were elected at the Annual General Meeting of the Society, held on 29th April; they will take office when the new session begins in the autumn:

Chairman
H. Edwards

Vice-Chairman
S. McLintock

Treasurer
S. Andrew

Auditor
L. S. Taylor

Honorary Secretary
S. Weir

Assistant Secretary
E. D. Harry

Committee

M. A. Hall
H. H. Johnson
R. Mayorcas
G. Millington
A. D. Osborne
R. Rippon
I. O. V. Thomas
R. L. Willott

EDUCATION

Non-destructive testing

An evening course of ten lectures on the techniques of non-destructive testing will be held in the Physics Department of the Brunel College of Technology, Woodlands Avenue, Acton, London W.3, every Wednesday from 7 to 8.30 p.m., starting on 7th October, 1959. Lectures will be given by specialists in the subject and by members of the College staff. These will be of interest to inspectors and others engaged on the work, and also to graduates and holders of the H.N.C. in engineering and science subjects, who require an introduction to the methods used in non-destructive testing.

The fee for the course will be £1. Further details may be obtained from the Physics Department at the College.

FORTHCOMING CONFERENCES AND EXHIBITIONS

X-ray analytical methods

A conference on X-ray diffraction and spectrography is being held at the Imperial College of Science and Technology, London S.W.7, from 6th to 10th July. It is being organized by Research and Control Instruments Ltd. in conjunction with the College's Department of Chemistry.

The conference will discuss and evaluate modern techniques and instrumentation for diffractometry and fluorescent analysis. Modern equipment will be available for demonstrating the principles of these methods over a wide range of applications.

Full information and registration forms may be obtained from the Conference Secretary at Research and Control Instruments Ltd., Instrument House, 207 King's Cross Road, London W.C.1.

Clean air

The Diamond Jubilee International Clean Air Conference and Exhibition is being held at Seymour Hall, London W.1. from 20th to 23rd October. The Conference will be devoted to the practical action required for the progressive reduction or abolition of air pollution. Full details may be obtained from the National Society for Clean Air, Palace Chambers, Bridge Street, London S.W.1.

Mineral processing

The Institution of Mining and Metallurgy is responsible for organizing the Fifth International Mineral Processing Congress, to be held in London at Church House, Westminster, from 6th to 9th April, 1960. The Congress covers fundamental and applied research and development in the fields of mineral dressing and chemical processing, excluding smelting. Further information may be obtained from the Secretary, The Institution of Mining and Metallurgy, 44 Portland Place, London W.1.

Compressed air and hydraulics

The organizers of the First International Compressed Air and Hydraulics Exhibition, which is to be held at Olympia, London, from 25th to 29th April, 1960, have announced that they have absorbed the Pneumatics and Hydraulics for Industry Exhibition and Convention, which was to have been held at Alexandra Palace, London, in October, 1959.

Hydrometallurgy

A symposium on hydrometallurgy, entitled "*Wet processing of minerals and industrial products*" is being organized by the Adelaide Branch of the Australasian Institute of Mining and Metallurgy from 16th to 19th February, 1960. The technical programme will cover wet sizing processes, agitation, liquid–solid separation, and recovery processes. Those interested should write to The Aus.I.M.M. Symposium Secretary, c/o South Australian Department of Mines, Osman Place, West Thebarton, Adelaide, South Australia.

Heat transfer

The Institution of Mechanical Engineers in collaboration with the American Society of Mechanical Engineers, is planning to hold a conference on heat transfer, in the USA in August, 1961, and in the UK in January, 1962. About 100 short papers on all aspects of heat transfer will be presented and discussed on both sides of the Atlantic. A British Committee has been set up to obtain papers from the UK and Europe; its chairman is Professor O. A. Saunders, of the Imperial College of Science and Technology. Further details will be announced in due course.

NEWS OF SCIENCE AND INDUSTRY

New blast-furnace at Margam

On 8th May Mr Harald Peake, Chairman of The Steel Company of Wales Ltd., performed the lighting-up ceremony which brought into operation the new No. 5 blast-furnace at Margam. The furnace is one of the main items of plant to be installed under the Company's £52 million development Scheme *M*.

The furnace, with a hearth diameter of 31 ft and a working volume of 59 365 ft³, is one of the largest blast furnaces in the world. It has an ironmaking capacity of more than 10 000 tons per week. It stands alongside No. 4 furnace which started up in January, 1956. The stockyard that serves No. 4 has been extended to cover both furnaces and now has a total capacity of 250 000 tons of iron ore. As the weekly requirements of foreign ore will be about 63 500 tons per week a fifth transporter unloader has been installed at Margam Wharf.

An important feature of the new furnace is that the layout makes it possible to operate entirely on sinter if this is required in the future. For this purpose a conveyor system links the ore-preparation plant direct with the sinter bunkers at No. 5 blast-furnace.

Air for the furnace is provided by a 125 000 ft³/min axial-flow turbo-blower of Swiss construction. To preheat the air there are three 140-ft high stoves each fitted with automatic valve change. The stove shells, furnace top, gas-cleaning plant, and mains have been designed for possible high top pressure operation.

When the furnace is tapped iron will be run into 180-ton capacity torpedo mixer ladles, each 55½ ft long. Slag will be run straight into pits alongside the furnace. From these pits it will be loaded by mechanical digging equipment into heavy road vehicles.

The furnace is carbon-lined to lintel level, and in the hearth has carbon sidewalls with a refractory centre; 2 700 tons of steel were used for the construction of the furnace itself and 390 000 bricks went into its lining.

Since No. 4 blast-furnace started up in 1956, it has produced more than 1 392 000 tons of very high quality iron and is now due for a routine reline. It is therefore intended to take No. 4 out of operation for relining as soon as No. 5 settles down to production.

For this reason the 80 new coke ovens installed at Abbey Works in anticipation of five blast-furnace operation will not be brought into production until No. 4 furnace is relined.

Committee on fatigue of engineering structures

A committee has been set up by the Council for Scientific and Industrial Research to review current research needs on the fatigue of engineering structures and to survey existing research facilities. Professor Sir Alfred Pugsley of Bristol University has been appointed chairman.

The terms of reference of the committee are: "To review current research needs on fatigue of engineering structures (metal), to survey existing research facilities, and to make recommendations to the Department of Scientific and Industrial Research." The committee is interested in examining records of failures or accidents that might have been due to structural fatigue, even if fatigue was not recognized as the initial cause.

Cold strip mill for Brinsworth

As part of a planned extension of their resources for the production of medium width strip, Steel, Peech and Tozer have placed an order with Davy and United Engineering Co. Ltd. for a new cold strip mill for their Brinsworth plant.

Fed by the Brinsworth hot strip mill, which was brought into production some months ago, the new cold reduction mill, a 10½ in. and 26 in. × 22 in. 4-stand tandem, will be capable of rolling mild-steel and high-carbon strip from 4 to 18 in. wide at speeds up to 2 100 ft/min. It will be supplied complete with a full range of auxiliary equipment capable of the mechanical handling of coils of strip eventually weighing over 5 tons each, and will be equipped with electronic equipment capable of controlling the thickness of the rolled strip automatically (Automatic Gauge Control). The whole plant is due to go into service in the summer of 1960.

Heavy plate levellers

The Head Wrightson Machine Co. Ltd. have received an order from Davy and United Engineering Co. Ltd. for a roller leveller to handle cold plates up to ¾ in. thick × 10 ft 6 in. wide, and another to handle hot plates up to 1¼ in. thick × 10 ft 6 in. wide. These levellers will form part of the new plate mill which is to be installed at the works of the Consett Iron Co. Ltd.

In addition, orders have recently been received from The South Durham Steel and Iron Co. Ltd. for two roller levellers to be installed in the plate mill of their new south works at West Hartlepool. Both of these roller levellers will handle plates in the cold condition. One has a maximum capacity of ¾ in. thick × 12 ft wide and the other will handle plates up to 1¾ in. × 11ft 6 in. wide.

All of these machines are of the backed up type and are designed to level plates in one pass over a wide thickness range.

Information conference in Sheffield

An Information Conference for the Sheffield steel industry was held on 23rd April under the chairmanship of Mr W. D. Pugh, Managing Director of the English Steel Corporation Ltd.

The conference was sponsored jointly by the Sheffield City Library, The Sheffield Chamber of Commerce, The Iron and Steel Institute, The British Steel Castings Research Association, The British Iron and Steel Research Association, and the British Iron and Steel Federation. It was designed to tell senior managers about the resources of information at their disposal, and to explain how they could be used to better advantage.

Mr C. Hurst of Samuel Osborn Ltd. and Mr. G. Parkin of Mappin and Webb Ltd. spoke, from the point of view of users, of the value of information in industry. Mr Hurst pointed out that the enormous store of existing technical knowledge was as valuable to a company as its own research department. Knowledge of patent literature was most valuable and enabled companies to avoid expensive infringements.

Mr Parkin said that an efficient information service, with its own staff, was of proved economic value to a firm large enough to establish one, but the need to use existing information was just as great in a small firm whose resources might not enable it to maintain its own information department. In this case the company should use outside experts as consultants; the cost was negligible in relation to the value obtained. Managers should make their staff "information minded" and realise that information could be as valuable as a whole battery of machines. The practical way to do this in a company that could not establish its own information department was to make one person responsible for these outside contacts.

Three papers were then given on sources of information. Mr M. L. Pearl, Librarian and Head of the Information Department of The Iron and Steel Institute, spoke about sources of technical information. He emphasized the great variety of such sources and the need to use those available inside an organization as well as those outside. An information officer needs to be aware of company policy and of the interests of his colleagues, as well as to be in touch with other organizations and experts. Companies planning to set up information departments and libraries could now obtain valuable advice from such bodies as ASLIB, the Sheffield City Library, and BISRA.

Dr E. Brewin, Head of the Intelligence Department of the British Iron and Steel Federation, spoke on sources of commercial information. He paid especial tribute to the BBC monitoring service in supplying information from abroad, and of the use of foreign embassies, consulates, and trade organizations. The main source of commercial information in this country was the daily and technical and trade press. The Federation had kept information on iron and steel companies in this country and abroad since 1934, including costs and prices, development, plant, etc., all of which was freely available.

Mr Bebbington finally described the special facilities available in the Sheffield City Library, which has outstanding collections on atomic energy and holds every British Patent from No. 1 issued in the year 1617. He mentioned SINTO—the Sheffield interchange organization to which over 40 local companies belong—and invited any company holding more than 50 technical books and subscribing to ten technical journals to join.

New rolling mill at Abbey

Another main item of the £52 million development Scheme *M* of The Steel Company of Wales Ltd. has come into operation with the start up of a new 4-stand cold reduction mill at Abbey Works.

The mill rolls steel strip for processing into thin sheets for sale as galvanized or black plates or for processing as tinplates. By filling a gap in the range of gauges supplied by the existing 3-stand and 5-stand cold mills, the new installation makes the Company's rolling capacity even more flexible.

To provide space for the new mill, an extension was made to the cold mill building so that the existing roll shop could be moved from its former location into the extension. This provided room for the installation of the 4-stand mill back-to-back with the 3-stand mill, with the motor-generator sets and electrical controls positioned between the two mills.

The new mill, which is driven by motors with a total horsepower of 19 750, is fed with steel coils from the pickle lines by means of an extension to the existing conveyor in the pickled coil store. After cold reduction on the mill, coils are conveyed by ram truck to the existing conveyor feeding the annealing department.

There are 21-in. dia. work rolls and 53-in. dia. back-up rolls, all of which are 56 in. long on the barrel, the strip handled being up to 50 in. wide. Strip thickness will normally vary from 0·120 to 0·076 in. on entering the mill and from 0·035 to 0·012 in. on the outgoing side. Coils up to 56 000 lb in weight will be rolled at speeds up to 3 500 ft/min.

The mill includes a number of unusual features, such as dihedral couplings on the main drives, new coil feeding arrangements, a new design of indicator for bearing lubrication failures, and improved hydraulic systems. A diversity of different roll coolants can be used.

Glossary of microscopy

The Royal Microscopical Society is compiling a glossary of the terms of microscopy, including electron microscopy. Members are invited to submit terms which they regard as requiring classification, with or without definitions, and with or without their equivalents in foreign languages, to Dr J. G. Cockburn at The Iron and Steel Institute. All contributions will be acknowledged in the *Journal of the Royal Microscopical Society*.

Corrigenda

On p. 78 of the paper by Levaux and Nepper (*Journal*, 1959, May, vol. 192), there are errors in the ordinates of three of the graphs. In Fig. 1 the ordinate should read 0–5, instead of 1–6; in Fig. 2 it should read 1–10; and in Fig. 3 it should read 0–10.

The heading of the fifth column of Table VIIIa in the paper by Durrer and Heintze (*Journal*, 1959, May, vol. 192, p. 19) should read "Ore, kg" and not "Ore, %" as printed.

On Fig. B of the "*Discussion on vacuum steelmaking*" (*Journal*, 1959, May, vol. 192, p. 71), the legend inside the graph should read "Tested at 20° C."

DIARY

6th–10th July—The Iron and Steel Institute—Special Meeting on the North-East Coast—Middlesbrough etc.

BRITISH IRON AND STEEL INDUSTRY TRANSLATION SERVICE

The following translations are now available, in addition to those given on page 174 of the June, 1959, issue of the *Journal*:

922. Eckinger, K.: "Modern Basis for the Development of Crane Bridges." *Stahl u. Eisen*, 1957, Jan. 10th, pp. 15–23, (£5)

1080. Engell, H. J.: "The Concentration Gradient of Iron-ion-vacancies in Wüstite Scaling Films and the Mechanism of Oxidation of Iron." (From German). *Acta Metallurgica*, 1958, June, pp. 439–445. (£3)

1163. Ludemann, K. F.: "An Economic Comparison of Various Steel Production Processes with Special Regard to the Development of Ferrous Metallurgy in the German Democratic Republic." *Neue Hütte*, 1958, Aug., pp. 462–474. (£6 10*s*. 0*d*.)

1194. Kornfel'd, V. N.: "Efficiency of Intensified Open-hearth Melting with Oxygen under Suitable Operating Conditions." *Stal'*, 1958, Dec., pp. 1095–1102. (£6)

1208. Heischkeil, W., *et al.*: "New Plant in the Basic Converter and Open-hearth Steel Plants of an Integrated Iron and Steel Works." *Stahl u. Eisen*, 1959, Jan. 8th, pp. 8–17. (£6)

1228. Voigt, H., and G. Schmalhaus: "Influence of Regenerator Design on the Pre-heating of Air and Gas in Triple-gas Fired Open-hearth Tilting Furnaces." *Stahl u. Eisen*, 1959, Jan. 22nd, pp. 80–88. (£5 10*s*. 0*d*.)

1230. Iweins, R.: "The Manufacture of Cast Steel Rolls at the Ougrée-Marihaye Works." *Rev. Univ. Mines*, 1945, Sept. 15th, pp. 122–128. (£4 10*s*. 0*d*.)

1238. Eichenauer, W., *et al.*: "Diffusion and Solubility of Hydrogen in α-iron and Silver." *Z. Metallk.*, 1958, May, pp. 220–225. (£3 10*s*. 0*d*.)

1241. Norén, T. M., and L. Kindbom: "Structure of Nitrided Heat-treatable Steels." *Stahl u. Eisen*, 1958, Dec. 25th, pp. 1881–1891. (£8 5*s*. 0*d*)

1245. Eicken, H. and W. Heidenhain: "The Effect of Drawing Speed, Reduction in Cross Section and Carbon Content on the Mechanical Properties of Drawn Steel Wires; Power Requirement in Drawing." *Stahl u. Eisen*, 1924, Dec. 18th, pp. 1687–1694. (£3 10*s*. 0*d*.)

1247. Ewers, J.: "Olefine- and Benzole-formation in Coke Ovens." *Brennstoff-Chemie*, 1955, Feb. 9th, pp. 33–37. (£3 15*s*. 0*d*.)

1250. Rosanow, W. B.: "Experimental Investigation of Pump Drive without Accumulators for Hydraulic Presses." From the German translation of the Russian book 'Hydraulic Presses,' ed. by A. I. Simin, pp. 130–141. (£3 5*s*. 0*d*.)

1252. Koch, K.-H.: "Development and Tasks of the Water Supply Service in the Iron and Steel Industry." *Stahl u. Eisen*, 1959, Jan. 22nd, pp. 89–94, (£4 5*s*. 0*d*.)

1255. Naeser, G., and W. Scholz: "Nitrogen Absorption during Air-refining." *Stahl u. Eisen*, 1959, Feb. 5th, pp. 137–141, (£3 10*s*. 0*d*.)

1257. Gondo, H., *et alius*: "Some factors which affect the Notch Impact Properties of Carbon Steels: the Influence of Aluminium and Nitrogen on the Strain-ageing of Steels." *Tetsu to Hagane*, 1957, Sept., pp. 894–897. (£5)

1262. Smirnov, V. S.: "The Occurrence of Internal Fracture during Cross Forging." *Vestnik Mashin.*, 1955, March, pp. 49–53. (£4 5*s*. 0*d*.)

1266. Kazantsev, V. A.: "Investigation of the K_β Group of the X-Ray Spectrum of Iron and Chromium in the Alpha and Sigma Phases of the Iron–Chromium System." *Dok. Akad. Nauk S.S.S.R.*, 1955, v. 101 (3), pp. 477–478. (£2)

1267. Rybnikov, V. A.: "Properties of Unfired Chrome–Magnesite and Magnesite–Chromite Bricks." (From Russian). *Ogneupory*, 1958, (7), pp. 303–307. (£2)

1268. Okhrimovich, B. P., *et al.*: "The Use of Unburnt Magnesite–Chrome Bricks for Furnace Roofs." *Stal'*, 1958, Feb., pp. 126–130. (£3 5*s*. 0*d*.)

1271. Brunn, F.: "A Contribution to the Problem of Pin-hole Porosity (Subcutaneous Blowholes) in Steel Castings." *Giessereitechnik*, 1958, Dec., pp. 307–310. (£3 5*s*. 0*d*.)

1275. Dahl, W., and W. Lueg: "Study of the Effect of the Rolling Conditions on the Properties of Hot-rolled Low Carbon Steel Strip. II—Mechanical Properties and Structure." *Stahl u. Eisen*, 1958, Dec. 11th, pp. 1793–1798. (£4)

1276. Schmitt-Thomas, K. G., *et al.*: "The Effect of Carbon Content, Cold Working and Heat-treatment on the Properties of Aluminium-Killed Deep Drawing Steel Strip." *Stahl u. Eisen*, 1958, Dec. 11th, pp. 1798–1808. (£5 5*s*. 0*d*.)

1284. Langer, N. A., and T. P. Novikov: "A Method for the Rapid Analysis of Flux AN–348–A." *Avtom. Svarka*, 1956, (5), pp. 80–83. (£2)

1307. Tepohl, W.: "Recent Experiences in One-coat White Enamelling." (Abridged). *Stahl u. Eisen*, 1959, Mar. 19th pp. 356–361. (£2 10*s*. 0*d*)

1315. Eckhard, S.: "The Spectro-chemical Determination of Phosphorus in Steel with the Phosphorus Line 3175 Å." (With discussion). *Arch. Eisenhütt.*, 1958, Feb., pp. 89–94. (£5 5*s*. 0*d*.)

1328. Perlin, I. L.: "Considerations of Stability in the Floating Plug Drawing Process for Tubes." *Tsvetnye Metally*, 1958, Sept., pp. 58–61. (£2 15*s*. 0*d*.)

ABSTRACTS OF CURRENT LITERATURE AND BOOK NOTICES

Iron and Steel Manufacture and Related Subjects

CONTENTS

Note: The letters, *a*, *b*, and *c* refer to the three columns on the page concerned.

	PAGE		PAGE		PAGE
Blast-Furnace Practice and Production of Pig Iron	301*a*	Heat-Treatment and Heat-Treatment Furnaces	306*a*	Protective Coatings	310*b*
Production of Steel	302*a*	Rolling Mill Practice	308*a*	Properties and Tests	311*a*
Foundry Practice	303*b*	Welding and Flame-Cutting ..	308*c*	Metallography	321*c*
		Machining and Machinability ..	310*b*	Corrosion	323*a*
				Analysis	323*c*

*Abstracts, the reference to which is marked with an asterisk are, or are to be made available as translations.

BLAST-FURNACE PRACTICE AND PRODUCTION OF PIG IRON

On Calculating Indirect Reduction in High- and Low-Shaft Furnaces. G. von Struve and R. Ebert. (*Neue Hütte*, 1958, **3**, Oct., 602–607). Laboratory tests carried out at Freiberg Mining College showed that CO_2 from iron carbonates is not decomposed in normal industrial operation, but that the dissociation of limestone is delayed, so that the CO_2 evolved reacts with solid carbon in parallel with the direct reduction of the oxides. A new formula for calculating indirect reduction has been derived from this observation.

A Contribution to the Study of Thin-Walled and Thick-Walled Blast-Furnaces. J. Libricky. (*United Nations, E/ECE* 305; *E/ECE/Steel*/119, 1958, 49–56; *Mineria Met.* 1958, **18**, Aug., 31–34). Work in Czechoslovakia on the thin-walled furnace and its operating characteristics and its advantages over conventional installations is reported.

Recent Advances in the Direct Reduction of Iron Ores. J. Astier (57–65). A review of processes aimed at the elimination of the blast furnace and advances in 1956, notably cyclo-steel, H-iron, the Krupp Renn process and the electric low-shaft furnace is given.

Screening Materials for the Blast Furnace. (*BISRA Summary* 127, 1958, pp. 4). See *J.I.S.I.* 1958, **190**, Sept., 55.

Operational Conditions of Blast Furnaces at Kuznetsk. N. N. Chernov. (*Metallurg*, 1958, (9), 4–6). These furnaces mostly produce low-Mn conversion iron with 0·76%Si, 0·53%Mn, 0·049%S and 0·15%P from a 75% sinter burden of basicity 1·4 and containing 51–52%Fe. Limestone consumption is ~120 lb/t, top pressure 0·65 atm and coke rate ~11 cwt/t, blast temp. 820–850° C and humidity 25 g/m³.—R. S.

A Study of the Movement of Materials in the Blast Furnace by Means of Radioactive Isotopes. G. G. Lukashev, Ya. S. Gorbanev, L. D. Prikhod'ko and D. V. Gulyga. (*Stal'*, 1958, (8), 682–687). Stack materials move most slowly at the walls and most rapidly in the combustion zone at the tuyeres, rates across the furnace vary. The maximum vertical rate, up to 33 ft/h, occurs at the top and this falls to ~16 ft/h at tuyere level. The tracer study did not give the changes in the direction of movement from the vertical. The differential rates of limestone and coke are discussed, also optimum rates. A. A. Cherepivskii and A. M. Skrebtsov. (687–690). The diffusion processes of isotope distribution in the metal at the hearth is discussed.

Study of the Descent of the Blast Furnace Burden. A. Kohn. (*Centre Doc. Sidér. Circ.*, 1958, (4), 817–818). A brief account of the use of radioactive Au and La for studying the rate of burden descent and of the mixing of iron and slag in the hearth is given.—B. G. B.

Processes Occurring in the Bosh of the Blast-Furnace when Melting with Oxygen-Enriched Blast. I. P. Bardin and M. Ya. Ostroukhov. (*Izvest. Akad. Nauk. Otdel. Tekn.*, 1958, (2), 7–14). The processes occurring in the bosh depend on such factors as distribution of materials in the furnace, the descent of the charge, the preparation of the ore, the size of the oxidizing zone, the protrusion of the tuyeres etc. and the effect of the oxygen-enriched blast is twofold: firstly, a shortening of the time the materials remain in the furnace resulting in the descent of insufficiently calcined ore and secondly a curtailment of the combustion zone (at high oxygen (over 25–26%) concentration in the blast), which results in an increase in the gas currents in the peripheral zones.—L. H.

Production and Blast Furnace Smelting of Fluxed Sinter from Kerch Ores. I. V. Raspopov, Ya. S. Gorbanev and D. V. Gulyga. (*Stal'*, 1958, (8), 676–682). Using a 80% sinter – 20% ore burden, increase of sinter basicity to 0·6 gives an increase in productivity of 6%, a fall in coke consumption of 8%, a reduction in cost per t of 10 roubles and a fall in limestone consumption of 32%. Full details of ore composition, sinter production and smelting practice are given.

Smelting High-Grade Conversion Pig Iron. P. S. Soloshenko. (*Metallurg*, 1958, (10), 4–6). Experience in the production of high-grade pig iron on the basis of coke containing 1%S and iron ore containing 0·012%P is described. The use of constant humidity blast and smaller compartments in hot blast stoves has reduced the coke rate.—R. S.

Calculation for a Burden of Complex Sinter of Krivorozhskii Ores. D. A. Kissin. (*Stal'*, 1958, *supplement* "*Blast Furnace Production*," *Metallurgizdat*, 1958, 67–74). Preliminary calculations are carried out to establish the coefficients of solid yield, which calculations include the determination of the coefficients of all components of the burden such as Mn ores, dust in the top of furnace, limestone and lime, coke dust etc. Formulae are derived for each set of calculations. With the method suggested, the calculations for the burden can be completed in 3–4 h, even with unsuccessful selection of raw materials, and the accuracy of the results is fully sufficient for practical application.—L. H.

Rationalising Blast Furnace Charging. L. Ya. Shparber. (*Metallurg*, 1958, (9), 10–11). The use of tubular vibrating conveyors for transporting materials to the skip at Magnitogorsk was associated with certain difficulties and the development of a new method of transporting hot sinter is described. The main idea is the replacement of a large number of small bunkers with large capacity bunkers suitably located to facilitate automation.

New Design for Cooling Blast Furnaces. S. M. Andon'ev, G. A. Kudinov and O. V. Filip'ev. (*Stal'*, 1958, (9), 776–780). A design using a thin walled stack with plate coolers and fitted with compartments for building up linings in the bosh is shown. Hearth, bosh and tuyere cooling devices are described.—R. S.

Utilization of Mn in Ferro-Manganese when Operating the Blast Furnace with Oxygen Enrichment. A. N. Redko. (**Metallurg*, 1958, (9), 7–10). Trials to maintain a high rate of coke combustion in spite of variations in burden quality were made. Data for 6 years are analysed and data on Mn volatilization are tabulated. Losses were increased from 10·9 to 15·1% by a fall in slag basicity but reduced volatilization of Mn from 78–74%. Considerable saving was shown by smelting Fe–Mn on basic slags with a higher Si content in the iron.—R. S.

Injection of Oxygen into the Blast Furnace V. Rémond. (*Technique Moderne*, 1958, **50** Aug., 363–367). Published work is reviewed from which general conclusions on thermal balance are drawn.

The Use of Radioactive Elements for Studies in the Blast Furnace. A. Kohn. (*Centre Doc. Sidér. Circ.* 1958, (4), 807–816). A review is given of the principles of radioactive tracer methods and examples are given of their use in blast furnace research. The examples quoted include: detection of the wear of stack brickwork, gas transit times, and rate of burden descent.—B. G. B.

Blast Furnace Operated with High Top Pressure. K. Réthy. (*Koh. Lapok*, 1958, **13**, May-June, 277–289). A comprehensive review with detailed theory is given of the use of high top pressures in blast furnace operation.

Continuous Control of Top Gas Composition in Blast Furnaces. S. T. Pliskanovskii and N. N. Temnokhud. (*Metallurg*, 1958 (10), 7–9). The previous method of automatic analysis of top gas at Azovstal permitted only the determination of CO_2 and did not give a correct picture of the thermal state of the furnace. A new optical-acoustic gas analyser which determines the CO and CO_2 contents is described.—R. S.

Blast Furnace Plant Operated by Steam-Enriched Blast. G. Schmidt. (*Koh. Lapok*, 1958, **13**, May-June, 269–276). The author discusses the advantages of steam enriched blast in blast furnace operation, describing Hungarian practice.—P. K.

The Effect of Sintered Ore on Blast Furnace Operation. L. Marczis. (*Koh. Lapok*, 1958, **13,** May-June, 236–240). The author reports Hungarian experiences on the effect of agglomerated ore in blast furnace operation.

PRODUCTION OF STEEL

A Review of Steelmaking Processes. T. P. Colclough. (*JISI*, 1959, **192,** July, 201–215). [This issue].

Thermal Characteristics of the 500-t Open Hearth Furnace. V. S. Kocho, V. I. Grankovskii and E. A. Ploshchenko. (*Stal'*, 1958, (9), 782–788). Comparison with 250-t furnaces and heat balances are made. Optimal conditions for a tap-to-tap time of not more than 11·5 h were found. Fuel consumption per t steel was reduced to 125–108 kg. Heat losses are reduced with the larger furnace. Blowing air through the caissons improved thermal conditions and reduced the coefficient of excess air to 0·9–1·05. With heat inputs of 35–40 m Cal/h, 5000 m³ of air was needed. It is expected that heat exchange conditions in 800 t furnaces will be much the same and a reduction in losses should increase thermal efficiency.

Pressure Distribution in Open-Hearth Furnaces. V. Parma. (*Neue Hütte*, 1958, **3,** Sept., 550–553). It is concluded that each O.H. furnace must be equipped with its own automatic pressure control system, and that steps must be taken to ensure that there is proper flame regulation, equal pressure on both sides of the furnace, and correct adjustment of the pressure controller.—L. J. L.

Equilibrium of Sulphur Distribution between Metal and Slag in Open-Hearth Furnaces. L. A. Shvartsman, A. I. Osipov, V. F. Surov, M. L. Sazonov, S. A. Telesov and A. M. Ofengenden. (*Doklady A. N.*, 1958, **120,** (3), 599–601). Various factors influence the passage of sulphur from slag to metal. It is suggested that efficient desulphurization of steel can be attained by a combination of thermodynamic, kinetic and compositional factors. Lowering of silica content in slag, maintenance of fluidity of slag by additions, including those containing FeO, all these may be helpful.—S. I. T.

Stepping Up the Open-Hearth Process by Means of the Use of Oxygen. I. P. Bardin. (*Hutn. Listy*, 1958, **13,** (10), 863–868). A description is given of experience in the use of O_2 to increase O.H. furnace output at the Azovstal' works in S. Russia, where the P content of the ores and of the resultant pig-iron is particularly high.—M. T.

Deposits Found in the Checkerwork of Open-Hearth Furnaces When Charging Large Quantities of Scrap. O. Dürmann and G. Mahn. (*Stahl Eisen*, 1958, **78,** Aug. 21, 1176–1180). When charging large quantities of unsorted scrap to a O.H. furnace a rapid reduction in draught was observed with all its consequences. It was found that the checkerwork was blocked by deposits consisting of more than 56%ZnO, with sulphides and Pb compounds. A method for cleaning the checker at regular intervals is described. This is essentially a reduction process using gas, the deposits were reduced to a tolerable level and the life of lining and roof of the furnace substantially extended.—T. G.

Reactions in the Open Hearth Affecting Refractory Life. H. M. Kraner. (*Amer. Ceram. Soc. Bull.*, 1958, July, 312–316). Slags and the attack by iron and Mn oxides on refractories are discussed MgO and silica brick are discussed and the effects of small amounts of Al_2O_3 are noted.

Determining the Economic Size of Recuperators. J. Kozina. (*Paliva*, 1958, **38,** (6), 188–192). [In Czech]. The influence of such variables as the cost of fuel, the cost of recuperators and running costs of these, on the size and design of recuperators are considered. Several nomograms are given.—P. F.

Deoxidising and Alloying of Steels for Very Heavy Ingots. A. Děkanovský. (*Hutn. Listy*, 1958, **13,** (7), 577–586). Most suitable and economical methods of deoxidizing and alloying 3·5%Cr steel cast as 150–250 t ingots for forging are discussed on the basis of published material and the author's own experiments. The most suitable method of deoxidizing consists of diffusion deoxidation in the 200-t O.H. furnace, lasting 15 min. using 3 kg of coke per t of steel, followed by a preliminary deoxidation using SiMn with a low Si-content, also in the furnace, and finally completion of the deoxidation and alloying with liquid Cr–Mn–Si–Al in the furnace for 10 min.—P. F.

Quantity and Nature of Oxide Particles Suspended in Basic Open-Hearth Steel During Melting Down. K. Born and H. Wittstruck. (*Stahl Eisen*, 1958, **78,** Oct. 30, 1514–1525). A method was developed for taking samples during melting of basic O.H. steel suitable as anodes for the analysis of the steel. It was found that the quantity of oxides suspended in the melt is negligible and that the particles are very small (0·5–3μ diameter). it is therefore not necessary to eliminate these particles. Some are crystalline and some amorphous. Microanalysis showed that the oxides consist of 60–80%SiO_2, balance Al_2O_3 + (sometimes) heavy-metal oxides. Further studies showed that—under industrial conditions—these oxides cannot be eliminated since it takes too long a time for the deoxidation products to reach the surface of the melt.—T. G.

Modifications to the Fontana Open Hearth Precipitators. E. V. Akerlow. (*Iron Steel Eng.*, 1958, **35,** July, 97–103). A description is given of tests made on an electrostatic precipitator used for the control of O.H. dust emission, and of modifications made to the design.

Deoxidation of Open-Hearth Steels during Duplexing. R. Štefec and J. Košan. (*Hutn. Listy*, 1958, **13,** (5), 397–402). [In Czech]. The chemical characteristics of O.H. melts destined for duplexing, i.e. O.H. to electric furnace, are discussed. Experimental melts from charges containing different proportions of pig-iron and high-chromium scrap, weighing 5, 18 or 35 t, were deoxidized by different methods. The results showed that the standard method, in which ferrosilicon and Al are added on the runner and ferromanganese in the furnace, can be replaced by a more economical deoxidation taking place entirely outside the furnace. The manganese-boil arising from additions of ferromanganese to the furnace was established not to have a significant effect on the degree of degasification, the final oxygen content or the quality of the steel. An effective, easily applied method of deoxidation is that induced in the runner of the O.H. furnace, or in the ladle. It can also be induced in the electric furnace after filling with the melt in the non-deoxidized, unkilled state. The lower manganese content in the O.H. furnace present with this method of deoxidation was not found to have undesirable effects; a higher pig-iron to scrap ratio than usual should, however be used in the charge.—P. F.

Electric Steel Production. F. Sommer. (*V.D.I.Z.*, 1958, **100,** Aug. 1, 1053–1062). A general description is given of the various types of electric steelmaking furnaces in use, together with their development, general characteristics and the types of steel manufactured.—L. D. B.

Electric Furnace Steel Manufacturing. W. E. Lewis. (*Iron Steel Eng.*, 1958, **35,** March, 98–102). The article reviews recent arc furnace development and its advantage both technical and economic—under certain conditions. Reference is made to arc furnace operation in conjunction with continuous casting and to the use of oxygen.—M. D. J. B.

Experiments on the Desulphurisation of Krupp-Renn Pellets and Their Subsequent Use in Electric Furnaces. M. Wahlster and C. B. Schmidt. (*Techn. Mitt. Krupp*, 1958, **16,** June, 39–40). Krupp-Renn pellets were desulphurized by treating in a rotary furnace at 800–1080° C with lime and coke. 80–90% desulphurization was attained, which was improved by subsequent melting. Pellets up to 4 mm diameter were treated.—L. D. B.

Fifty Years of Electric Steel Production in Czechoslovakia. J. Korecký. (*Hutnik*, 1958, **8,** (6), 185–187). A brief historical survey of the development of the production of electric steel from its beginnings in Kladno in 1908.—P. F.

Electric Arc Steel Furnaces. F. N. Pensa. (*J. Four Elect.*, 1958, **63,** (4), 132–137). An outline is given of the development, structural features and performance of the modern electric arc steelmaking furnace; the author discusses the possibilities of the arc furnace as a tonnage unit.

Stirring in Steel Production. A Faerden. (*Metallurgia*, 1958, **58,** Aug., 57–63). The methods of stirring the arc furnace by mechanical and inductive means and by injection of gases or metal vapours, reladling and by rotation of the furnace are reviewed, and costs are compared. (33 refs.).

The Stirring of Metal in Electric Arc Furnaces. P. E. Hammarlund. (*Hutn. Listy*, 1958, **13,** (10), 868–872). A lecture describing the application and economic aspects of the use of stirring mechanisms in arc furnaces. It is considered that their use is only advantageous in large furnaces for the melting of special and alloy steels.—M. T.

The Rational Use of a Choke Coil in Arc Furnaces. J. Pietka and F. Fikus. (*Hutnik*, 1958, **25,** (7–8), 278–284). [In Polish]. The role of a choke coil in the circuit of an arc furnace is discussed. To achieve the best results the inductive reactance of the choke should be as small as is compatible with smooth running, restricting power oscillations and voltage oscillations in the grid. The results of experiments using a choke installed on an arc furnace in the Baildon plants are given. The rational use of the choke resulted in shortening the time of melting, reducing the expenditure of energy and increasing cos ϕ.

Possibility of Charging Arc Furnaces Above Their Capacity. A. I. Sapko and Z. I. Sapko. (*Metallurg*, 1958 (9), 20–21). It was shown to be irrational to overload the 30–40-t arc furnaces at Dnyeprospetsstal' by more than 30%. These furnaces have a bath diameter/bath depth ratio of 6·6. The right way to raise productivity is by improving the electrical features of these furnaces and by the use of oxygen.—R. S.

Production of Resistance Alloys in Electric Furnaces. A. F. Kablukovsky, Ya. S. Leizerov and I. P. Solodikhin. (*Metallurg*, 1958, (9), 12–15). The production of Nichrome-type alloys at the Elektrostal' plant is described. The former method of melting in 20-t basic arc furnaces using oxygen and adding Cr and Al during refining was replaced by melting in 5-t arc furnaces using ore and adding Cr and Al during refining, with prior deoxidation of the bath with Al and Ti. In the new method, carbon tetrachloride is used as a protection against oxidation during pouring. Structural soundness of the resistance alloys is greatly improved.—R. S.

Alloying Steels with Boron by Means of Borax. J. Skála and D. Tlustá. (*Hutn. Listy*, 1958, **13,** (5), 418–421). An account is given of a series of experiments in which it was established that by adding borax to steels together with the deoxidizing elements the steels are effectively and more economically alloyed with boron than on using ferroboron and other currently used alloy additions. Borax can also be used in cast irons, as these contain large amounts of strongly reducing elements, i.e. C and Si.—P. F.

Preliminary Note on the Making of Free Machining Steels Starting with Galena. V. Mazzarella. (*Bol. A.B.M.*, 1958, **14,** Apr., 153–180). [In Portuguese]. A review of the properties and uses of the free machining steels is given and followed by a description of experiments which attempted to produce leaded steels by adding galena to the melt. Melts were made in an induction furnace, the galena being added as powder of known particle size. The steels were made to a 0·30%C specification and examined for forgeability, machinability, mechanical properties and lead segregation. These preliminary tests show the possibility of adding 0·15% lead to steel by means of galena. Sulphur is also taken up to about 0·20%,

but does not affect the forgeability if it is balanced by manganese. (25 references).

The Production and Consumption of Cast Iron Ingot-Moulds. S. Hargitay. (*Koh. Lapok*, Öntöde, 1958, **9**, May-June, 101–109). The author discusses the various factors, which effect the life of an ingot mould such as preheating; time of ingot stripping; cooling; method, temperature and rate of steel teeming; steel composition; size, construction, structure and chemical composition of the ingot mould. Afterwards he describes the main principles of adequate cast iron ingot mould construction and production.—F. K.

Repair Welding of Steelworks Moulds. K. Petschauer. (*Giessereitechn.*, 1958, **4**, June, 125–128). A practical description with illustrations is given of the repair of ingot moulds, which have been eroded or cracked, by means of welding.—L. D. H.

Welded-Steel Teeming Ladle. P. M. Kichaev and G. A. Dubrovich. (*Vestnik Mashin.*, 1958, (8), 56–57). Theoretical considerations are explained and formulae and graphs given by means of which to calculate the dimensions of the ladle such as thickness of bottom, sides, etc.—L. H.

Platform Cuts Ladle Relining Costs. (*Steel*, 1958, **142**, June 30, 88). A safe and efficient method of lining ladles is described. Steps are: clean out ladle; lower in a preloaded pallet of bricks and line the bottom; lower in platform holding suitably stacked bricks in the middle and line the ladle to half height; lower in table and another platform from which to complete the lining.—D. L. C. P.

Metallurgical Problems Arising from the Teeming and Solidification of Rimming Steel. H. Rellermeyer and R. Hammer. (**Stahl Eisen*, 1958, **78**, Oct. 30, 1505–1513). The distribution of P, S, N, C, and O was studied in ingots of 3·8–7 t of rimming basic Bessemer steel and in the stock after rolling. The effects of [C] and [O] and their ratio on solidification and segregation are discussed and the findings of structure investigations reported with particular emphasis on the segregation of S and P. The O content of the ingot is determined mainly by the C level of the steel. No correlation between O content and ingot size was found.—T. G.

The Reduction of the Shrinkage Cavity in Steel Ingots by Application of Exothermic Linings. R. Frackiewicz. (*Hutnik*, 1958, **25**, (7–8), 262–268). [In Polish]. The author describes a method patented by himself for reducing the size of the shrinkage cavity in sink heads for steel ingot moulds. The method consists in providing the interior of a sink head inside the refractory lining with an exothermic lining. This lining is made of a mixture of Al–Mg alloy powder, Swedish ore and some other components. Tests have been carried out in Warsaw and Baildon steel plants; about 150 t of low alloy steel and 80 t of stainless steel were used. The weight of ingot varied from 110 to 1060 kg. Considerable reduction in size of the shrinkage cavities was obtained; economy of steel in some cases was 16%, on the average 8%. Reduction of costs is appreciable, particularly for high alloy steels. The article is illustrated with several macroscopic sections of ingots.

The Problem of Casting Steel for Transformers. M. Stankiewicz. (*Hutnik*, 1957, **24**, July-Aug., 266–271). [In Polish]. Chemical composition and the effect of various elements on the magnetic properties of steel for transformers are discussed. Methods of casting this type of steel from electric and O.H. furnaces are described. The following recommendations are made: (1) Si contents should be 4·3–4·5%; (2) Electric arc furnace should be used in preference to O.H.; (3) Further improvements of steels are suggested by casting in vacuo.

Designing a Large Tonnage Continuous Casting Plant. H. E. Skelley and R. Easton. (*Iron Steel Eng.*, 1958, **35**, May, 131–142). This paper summarizes factors which must be considered when contemplating the construction of a continuous casting plant. It evaluates the advantages and disadvantages of the various types of plant and lists the merits of continuous casting over rolling. (11 References).—M. D. J. B.

FOUNDRY PRACTICE

Shooting Borings into Cupola Reduces Melting Cost. R. E. Dixon. (*Foundry*, 1958, **86**, May, 218–221). Equipment for injecting cast iron borings into the cupola melting zone is described. Less coke is needed to melt borings because heat is transferred to them more efficiently than to large-size cast scrap. Borings melt within a few seconds after contact with incandescent coke. The borings constitute 5–10% of the charge and are injected at batches of 7–9 lb every 10 sec by compressed air. The price of borings is 25–30% lower than that of No. 1 cupola scrap.—T. G.

The Use of Calcium Carbide in the Cupola. R. Schulze. (*Giesserei*, 1958, **45**, Sept. 11, 542–545). Calcium carbide is a secondary fuel whose effect is due to the high temperature of the combustion zone which is displaced downward. Carbide melts at 1650° and drips down into the furnace where it is burnt with oxygen. The heat released increases the temperature above the nozzles and promotes coke combustion. A maximum temperature zone (1750–1800°) is created just above the tuyere level. The heat content of the coke is more thoroughly exploited: coke consumption is reduced by 25%.—L. J. L.

Oxygen-Injection Process Aids Cupola Combustion Control. J. R. La Pota. (*Foundry*, 1958, **86**, May, 269–276). A new method of using O_2 in the cupola called the "controlled cupola chemistry process" is described. Streams of pure O_2 are injected directly into each tuyere at a velocity somewhat greater than that of the air blast, thus creating a localized concentration of O_2 in the combustion area. Advantages and applications of the process are listed.—T. G.

The CTIF-Ulmer-Recuperator for the Hot Blast Cupola. G. Ulmer. (*Giesserei*, 1958, **45**, Sept. 11, 531–541). In the novel design described, a twin-walled recuperator takes the place of the cupola chimney immediately above the furnace throat. Cupola waste gases flow through the cylindrical centre space, and the heating air flows between the two cylinder walls, in an upward direction in both cases. Results of tests on experimental and industrial cupolas are reported.

A Regenerative Reverberatory Foundry Furnace Fired with Coke-Oven Gas. G. Knaginin and R. Andrzejewski. (*Hutnik*, 1958, **25**, April-May, 165–171). The construction of this 25-t furnace is fully described with drawings, operational data, heat balances etc. The thermal efficiency figures are shown to be higher than those of comparable furnaces.

A Modern Melting Plant for Cast Iron. V. Frey. (*Giesserei*, 1958, **45**, July 3, 397–402). The modern foundry constructed by Escher Wyss AG is described. It consists of three mains frequency furnaces, one electric arc furnace and two cupolas, and is highly mechanized.

The Héroult Type of Acid Electric Furnace in the Steel Foundry. Viallon. (*Usine Nouv.*, 1958, **14**, Oct. 2, 23). A summary of a paper presented at a conference of the Association technique de la Fonderie in May on the construction and operation of the acid lined 3-phase furnace with a report on the productivity of a 2 tonne unit at the Acéries d'Hirson.

Demag Electric Furnaces in Steel Foundries. R. Genwo. (*Demag Nachrichten*, 1958, (151), 7–13). The development and performance of acid and basic Demag electric arc furnaces are described, the types of steel manufactured enumerated and costs for basic and acid electric are practice, acid Bessemer and medium frequency acid lined crucible operation are compared.

Induction Heating of Light and Heavy Metals and of Steel. W. Koester. (*Z. Metallk.*, 1958, **49**, Oct., 523–526). The theory and degree of efficiency of induction heating installations are discussed; the construction and mode of operation of such plants are described; two methods of achieving a given degree of uniform radial temperature are reported.—L. J. L.

Uses of the Mains Frequency Coreless Induction Melting Furnace. L. G. N. Edwards. (*Iron Steel Rev.*, 1958, **1**, Feb., 53–57). The principles and characteristics of the mains-frequency furnace are summarized and its application to the melting of steel, iron and non-ferrous alloys is discussed.—B. C. W.

The Accuracy of the Term "Modified Cast Iron." J. Piaskowski. (*Prz. Odlew.*, 1958, **8**, June, 169–172). The author discusses definitions of "modified cast iron" which have appeared in Polish publications and concludes that it lacks precision.—M. T.

Spheroidal Graphite Cast Iron. (*Metalen*, 1958, **13**, May 20, 92–97). [In English]. A general description of the properties and some examples of the applications of nodular cast iron.

Spheroidal Graphitic Cast Iron. J. Cournot. (*Mét. Corr. Ind.*, 1958, **33**, May, 191–192). An introductory note.

Theoretical Considerations in Relation to the Formation of Spheroidal Graphite. M. Ferry. (*Mét. Corr. Ind.*, 1958, **33**, May, 193–211). A discussion and critical comparison of theories relating to the formation of nodular and lamellar graphite in cast iron. (62 refs.)

The Manufacture and Properties of Spheroidal Graphitic Cast Iron. M. Eudier and L. Hallot. (*Mét. Corr. Ind.*, 1958, **33**, May, 212–227). An outline of the chemical composition of the iron, the various Mg alloys used for inoculation and methods of introduction, is followed by a summary of the physical properties of nodular cast iron.

Investigation of Nodular Cast Iron Manufactured in the Springfield Armory Casting Laboratory. O. Lubinkowski. (*U.S. Dept. Commerce*, P.B. 131399, 1957, May 28, pp. 45: SA–TR18–1049). Thin walled castings (below ¼ in.) using Mg–Si as inoculant were made with hypo- and hypereutectic irons, and heat treatments, analysis and physical tests are reported. Fe–Mn, Te, Lanceramp and Fe–Si were also evaluated as inoculants.

An Austenitic Manganese Nodular Cast Iron. N. Popescu. (*Studii şi Cercetări de Metalurgie*, 1958, **3**, (1), 33–38). [In Rumanian]. Data are provided on the manufacture, chemical composition and mechanical and physical properties of an austenitic manganese steel with a spheroidal graphitic structure; these properties are compared with an iron of similar chemical composition, but with lamellar graphite.

Weight and Thermal Balance for the Production of Spheroidal Quality Refined Iron in the Electric Furnace. G. Zuliani. (*J. Four Elect.*, 1958, **63**, (4), 129–131). A supplement to a previous article (cf. *ibid*, 1957, (2).) On the manufacture in the electric furnace of iron intended for the production of nodular cast iron.

Effect of Ferrosilicon and Magnesium Inoculation on Nuclei Formation in Cast Iron. F. Hurum. (*Trans. A.F.S.*, 1957, **65**, 66–74). Trials with a special experimental furnace are reported, and a principle of fractional diffusion preciously advanced appeared to be confirmed. ε-, α- and γ-phases were seen to be formed successively. When Mg was added, silicon carbides were seen to form inside undissolved Fe–Si alloy and C diffused into silico-austenite. The formation and nature of black X-particles is considered.

Spheroidal Graphite in Sweden. M. Pachowski. (*Prz. Odlew.*, 1958, **8**, July, 204–208). A description is given of the two methods of producing S.G. iron most commonly adopted in Swedish foundries, as practised at A.B. Bjorneborgs Jernverk and A.B. Svenska Kullagerfabriken, Katrineholm. At the former, inoculation is effected by using an alloy of 15%Mg + 85%Ni, and at the latter with a Mg-rich alloy (about 90%).

Titanium in Nodular Cast Iron. A. Tominaga, I. Aoki, T. Tottori. (*Metal Progress*, 1958, **74**, Aug., 116). The results of an investigation into the effect of tramp elements on the

formation of graphite nodules in ductile cast iron are given. Ti, in combination with certain other elements inhibits nodule formation. The other elements in order of decreasing effect are Bi, Pb, Sb, As, Sn.—D. L. C. P.

Acid-Resistant, High-Silicon Nodular Cast Irons. T. R. Dumitrescu, V. Medeleanu, M. Nicolaid and I. Dinu. (*Studii si Cercetări de Metalurgie*, 1958, **3**, (1), 7–17). [In Rumanian]. The structure, mechanical properties and resistance to corrosion of high-silicon irons containing 8 to 24%Si with lamellar and spheroidal graphite respectively are compared. Corrosion tests were carried out with 20%HNO_3, 30%H_2SO_4 and 10%HCl at boiling point; the nodular structure was obtained by the addition of Mg.

Alloying and Inoculation of Heat-Resisting Silicon Iron with Magnesium. P. I. Durasov and N. N. Aleksandrov. (*Metall. Obrabotka Met.*, 1957, (5), 42–48). The alloying of silicon irons containing laminar graphite with carbide-forming elements increases growth. It is not expedient with silicon irons containing nodular graphite as it lessens immunity from growth and reduces fluidity. The alloying of silicon irons with Al however, increases their resistance to scaling though this is accompanied by an increase in brittleness and deterioration of the casting properties together with a lowering of the resistance to growth. The iron possessing the greatest heat resistance of all those investigated was the unalloyed silicon iron with a nodular graphite structure and containing 5–6%Si.—R. S.

Blackheart and Whiteheart Malleable Cast Irons. M. Gailly. (*Rev. Gén. Méc.*, 1958, **42**, May, 283–287). A brief outline of the properties and uses of blackheart and whiteheart malleable cast irons.

Foundry Applications of the Calcium Carbide Injection Process. W. R. Lysobey and A. E. Tull. (*Mod. Castings*, 1958, **34**, July, 53–61). The technique of the injection of CaC_2 to produce flake graphitic irons is described, and the results of trials are discussed. In addition to reducing S content, the addition of CaC_2 lowers the chilling tendency of the iron and promotes the formation of type A graphite in a pearlitic matrix, thus removing many difficulties associated with porosity and shrinkage.

Pearlitic Malleable Manufacture and Use. O. K. Hunsaker. (*A.S.M.E.* 1957, 57–A–241, pp. 5). Manufacture and control, physical and heat-treating properties of the material are reviewed as well as uses, availability and relative economy.

Ductile Iron—Alloyed and Normalized. C. R. Isleib and R. E. Savage. (*Trans. A.F.S.*, 1957, **65**, 75–83). Unalloyed, Ni-containing, Ni–Mo-containing and Ni–Mo–V-containing irons were examined as cast, normalized and normalized and tempered using different section sizes. Hardness, tensile properties, structures, elongation and impact strengths were measured and are plotted against composition and treatment so as to provide a guide for choice of materials.

Investigation of the Effect of Processing Variables on Mechanical Properties of Pearlitic Malleable Iron. H. H. Johnson and W. K. Bock. (*Trans. A.F.S.*, 1957, **65**, 450–461). Furnace atmosphere, austenitizing temperature, quenching rate, retempering temperature, tempering time and surface conditions were studied. Statistical study of the results showed significant correlations. Surface condition was highly and atmosphere and austenitizing temperature much less significant.

Alloying and Heat-Treating Ductile Cast Iron. (*U.S. Dept. Commerce, PB*, 131350; 1955, Aug. 31, pp. 89). The basic variables in the risering of ductile iron castings were determined and effects on shrinkage measured. Risering for gross shrinkage was accomplished by keeping the riser liquid as long as necessary while it contained adequate quantities of metal and was within feeding distance. The amount of metal needed depends on C and Si content, pouring temperature and mould rigidity. Sponge-type shrinkage was due to lack of graphite precipitation and was eliminated when %C + 1/7%Si = 3·9%. Machinability was then considered and the forces involved in cutting ductile iron and the effects of cutting fluids on these forces and on surface finish was studied.

Heat-Resisting Cast Irons. A. Lévi. (*Fonderie*, 1958, Sept., 412–420). The influence of several alloying elements on the thermal properties of cast irons is discussed. Pearlite, cementite and steadite in cast iron are more resistant to corrosion by hot gases than the ferritic steels. Cast irons undergo irreversible expansion and volume increase on heating. Steels are more resistant than cast irons to thermal shock. Carbide stabilizers such as Cr improve the heat-resisting properties of cast iron. (25 references).—R. P.

Tin as an Alloy in Grey Cast Iron. J. A. Davis, D. E. Krause and H. W. Lownie, jun. (*Trans. A.F.S.*, 1957, **65**, 590–597). It is shown that the carbide-stabilizing effect of tin can be turned to advantage with careful control. Microstructures are shown and mechanical properties are determined. Contents of 0·1% or less improved the iron and properties fell off at higher contents. Hypereutectic irons contained some residual ferrite, but hypoeutectic automotive iron was fully pearlitic with 0·05–0·1%Sn.

New Foundry for Precision Casting Steel. (*Metallurgia*, 1958, **58**, July, 29–31, 36). A description is given of the new Holbrook foundry of Messrs. Samuel Osborn, constructed for precision casting by the shell moulding, Osborn-Shaw and CO_2-block processes.

Factors Affecting the Toughness of Mild Steel Castings. H. H. Fairfield and J. A. Ortiz. (*Mod. Castings*, 1958, **34**, July, 70–74). Factors influencing the toughness of mild steel castings, as measured by reduction in area, are considered for acid and basic electric arc melting practice. (16 refs.).

Combined Melting Processes for the Manufacture of Steel in Steel Foundries. S. J. Karmasin, W. M. Schestopal and J. L. Dodin. (*Giessereitechn.*, 1958, **4**, Aug., 192–194). The advantages of using various combinations of melting units are discussed. In the case of iron where close tolerances of Si, S and P are difficult to maintain, it has been found advantageous to use a basic cupola, followed by treatment in a side-blown converter with oxygen-enriched blast, with a final refining treatment in an acid-lined electric arc furnace.

Casting Ferrous Metals in Anodized Moulds. L. D. LaMotte. (*Prec. Met. Mold.*, 1958, **16**, Mar., 46–48). The use of anodized Al permanent moulds for the casting of stainless steel and other ferrous and non-ferrous metals is said to have certain advantages owing to their high thermal conductivity.

Problems of the Flowability of Moulding Sands. L. Lewandowski. (*Prz. Odlew.*, 1958, **8**, July, 201–204). Methods and criteria for the assessment of this factor are reviewed and discussed on the basis of foreign literature sources. The importance of good flowability and its affect on the quality of moulds are indicated.—M. T.

Effect of Temperature on the pH of Foundry Sands. N. D. Brinkmann and G. Gottschalk. (*Trans. A.F.S.*, 1957, **65**, 84–87). Castings containing the cores were sent through an annealing kiln at ~1700° F which burnt off the binder and allowed the sand to flow out freely. It then resembled new sand, but produced very weak cores. The pH was found to be 11 and this could be the cause of the reduction in strength. The sand alone also gave this result on heating so samples of various sands were similarly heated and their properties measured. Variations were found from sample to sample even from the same source. The effect is attributed to fossil $CaCO_3$.

Correlation Between Casting Surface and Hot Properties of Moulding Sands. A.F.S. Sand Division Committee. (*Trans. A.F.S.*, 1957, **65**, 128–133). A study of the relationship between expansion-type scab, buckle and rat-tail defects and moulding sand properties. Two simple ore-specimens tests were developed, one, a shock expansion test for mould wall expansion at 1800° F under constant load of 1 p.s.i. for 22·5 sec., and a restraining load test at 1800° F with a 5-min. soak. Methods of reducing the tendencies are given.

Influence of Sand Grains Distribution on Green-sand Casting Finish. C. E. McQuiston. (*Trans. A.F.S.*, 1957, **65**, 188–206). An experimental casting method and a surface finish criterion were devized and used to correlate finish with grain size. The best finishes were obtained with large grain fineness number and wide spread of distribution. The effect of squeeze pressure is small. (31 refs.).

Scabbing of Steel Moulding Sands. A.F.S. Sand Division Committee. (*Trans. A.F.S.*, 1957, **65**, 235–237). Two years' work is outlined. A restraining load test was used which correlated with test casting results for scab formation but total scab length could not be predicted. The loaded shock expansion test did not correlate. Increasing moisture produced scabbing from non-scabbing sands.

Properties of Moulding Materials at High Temperature. II. Expansion by Shock Heating. K. Futaki. (*Rep. Gov. Ind. Res. Inst.*, 1957, **6**, June, 297–305). Silica sand expands more on shock heating than by slow heating due to disintegration of grains and cracking in the test piece. Expansions under various moulding conditions are determined and related to casting defects. **III. Effects of Binders and Additions on Hot Strength and Deformation of Sand Moulds.** (Oct., 538–543). Bentonite and clay as binders and pitch, coal powder, wood flour, cereal, etc. as additions were examined for their effects. Bentonite-bound sands have a maximum strength at 900–1000° C and clay-bound sands at 1000–1100° C. Wood flour decreases hot strength and increases hot deformation. Pitch and powdered coal increase strength up to 2% and decrease it over 3%. Red ochre has no effect up to 5%.

Combined Permeability of Facing and Backing Sands and its Relationship to the Surface Roughness of Castings. K. Hashimoto. (*Rep. Gov. Ind. Res. Inst.*, 1957, **6**, May, 239–244). An equation is given for combined permeability. Over a range of values for combined permeability, surface roughness decreases to a minimum, and control in relation to sand grain size and mould surface smoothness is required.

The Influence of Height of Test-Piece on the Properties of Moulding Sands. K. Hashimoto (*Rep. Gov. Ind. Res. Inst.*, 1957, **6**, Sept., 485–489). Test-pieces from 25–75 mm were used and an equation for the relation to compressive strength is given. Permeability is not affected, top hardness increases linearly and ramming energy also increases proportionally to height.

Effect of Vent Wire on the Permeability of Moulding Sands. K. Hashimoto. (*Rep. Gov. Ind. Res. Inst.*, 1957, **6**, July, 357–359). The permeability ratio before and after insertion of vent wires is defined as improvement modulus of permeability. When the insertion depth is varied a relationship is formulated and similarly for number of insertions.

Surface Roughness of Iron Castings. O. Carlsson. (*Gjuteriet*, 1958, **48**, Aug., 111–115). The effects of sand grading and composition, binders, the properties of the moulding machine and the casting temperature on the surface roughness of the casting are discussed.

Cooling Plant for Dried Quartz Sand. K. Haake. (*Giessereitechn.*, 1958, **4**, June, 123–124). The construction and operation are described of a drying plant which cools rapidly and effectively the hot moulding sand from a drying oven. Dust is efficiently removed.—L. D. H.

An Investigation of a Moulding Mixture Based on Silica Sand with Coal Slurry. J. Szreniawski and E. Wozniak. (*Prz. Odlew.*, 1958, **8**, June, 153–156). The article comprises the results of research into the strength and permeability of two moulding mixtures prepared from typical fine and coarse grain silica sands, to each of which coal slurry was added as a bonding agent (C–39·62%, S–1·26% Ash–54·40%, of which SiO_2–56·5% and Al_2O_3 32·87%). After drying the sand and

the slurry, the slurry was pulverized to give a 30% residue on an AFS 100 screen, and up up 18% slurry added to the sands. Permeability, compression and tensile strengths of green and dry mixtures were measured on a Ridsdale type apparatus. Results show that such mixtures have a wide field of application for castings, as their properties fulfil the requirements of green and dry moulding mixtures in practice. It is also evident that better properties may be obtained from mixing coal slurry into coarse grained sand, especially on account of the high permeability of the latter. It remains to assess the behaviour of such mixtures at high temperatures.—M. T.

Physico-Chemical Properties of Spanish Clays. Mechanical Tests. P. de la Cierva and F. de Andrés. (*Rev. Ciencia Apl.*, 1958, **12**, 289–292). The compressive strength at high and low temperatures and the bending strength of Spanish clays are examined. The third of a series of articles.

Effect of Various Clays and of Tempering Method on Sand Properties and Casting Quality. A. E. Murton. (*Trans. A.F.S.*, 1957, **65**, 379–385). Four bonding clays were tested by methods simulating synthetic and natural bonded sand treatments. Castings were poured in these and rated for scabbing and surface finish. Western bentonite gave the best results. In the unmulled condition only illite gave a quickly-prepared and easily controlled mixture.

The Influence of Heating on the Strength of Waterglass Sand. H. G. Levenlink. (*Metalen*, 1958, **13**, Sept. 30, 326–333). This section describes the results of investigations into the influence of pressure, duration of CO_2 treatment, organic additions and the Na_2O/SiO_2 ratio on the strength of sodium silicate-bonded sands.

Evaluation of Bentonites in the Steel Foundry. V. E. Zang. (*Foundry*, 1958, **86**, May, 154–157). Testing methods for bentonites with particular emphasis on pH values are described, the minimum acceptable pH value being 8·2. Pictures of the surfaces of test castings from moulds made of sand mixes containing various amounts of bentonite are reproduced.

Investigations on the Effect of Heat on the Bonding Properties of Various Bentonites. F. Hofmann. (*Mod. Castings*, 1958, **34**, July, 31–37). An account is given of the development of comprehensive methods of testing bentonites for foundry use, making use of thermal analysis techniques. The results are discussed.

Moulding Sand Clays. II. Relationship between Bond Strength and the Area of Contact. T. Yoshihara. (*Rep. Gov. Ind. Res. Inst.*, 1958, **7**, Jan., 39–44). Glass particles were used bound with clay slurry. The strength was found to be proportional to contact area.

Trial Production of Moulding Binders from Tall Oil. S. Ozaki, K. Murata, H. Ikegami and R. Tatematsu. (*Rep. Gov. Ind. Res. Inst.*, 1957, **6**, May, 245–251). Esters with glycerol, pentaerythritol and mannitol, modified glyptal and maleic alkyd resins with or without a drying oil were used. The maleic alkyds prepared from mixtures of tall oil and drying oils were the best.

Oil-Bonded Moulding Sand. R. G. Megaw and K. A. Miericke. (*Trans. A.F.S.*, 1957, **65**, 513–516). Advantages of waterless mixes are given and the use of an organic derivative of bentonite known as Bentone is described. Stainless steel castings were made, but the products were mainly non-ferrous.

Wet Strength—a Novel Concept to Characterize the Behaviour in Practice of Clay-bound Moulding Sands, and the Effect of Activation on Wet Strength. W. Patterson and D. Boenisch. (*Giesserei*, 1958, **45**, Sept. 11, 565–567). A temperature gradient forms in the sand surface under the influence of the heat of casting; this leads to inhomogeneous moisture distribution. An excess moisture zone of reduced strength forms under the dry surface of the sand. The method described permits quantitative measurement of strength in this zone. It also permits quantitative measurement of the degree of activation in bentonites.

Surfacing Agents. J. Leonard. (*Fonderie Belge*, 1958, **28**, July, 206–208; Aug., 234–238). The first section deals with the characteristics and method of application of some casting surfacing agents. The second part discusses some difficulties, disadvantages and special cases, and concludes with a description of laboratory trials.

Foundry Patterns. S. S. Khanna. (*Iron Steel Rev.*, 1958, **1**, Feb., 91–94). The various types of pattern, the materials used in their construction, and the dimensional allowances necessary in their design are described.

Tisco Pattern Shop and Foundries. M. S. Dua. (*Iron Steel Rev.*, 1958, **1**, Feb., 63–67). A description is given of the foundry facilities at the works of the Tata Iron and Steel Co.

Saving Time. L. A. Dodge. (*Mod. Castings*, 1958, **34**, July, 22–23). The use of epoxy resins for the rapid and accurate duplication of foundry patterns is described with the aid of an illustrated example.

European Foundries and Pattern Shops. E. T. Kindt. (*Trans. A.F.S.*, 1957, **65**, 94–96). A report in general terms of visits to European (Holland, Belgium, Germany) foundries, also Esher Wyss in Switzerland, with special reference to patternmaking.

Synthetic Products as a New Material for Foundry Patterns. H. Wasowicz. (*Prz. Odlewc*, 1958, **8**, June, 157–163). Plastic materials for the preparation of patterns, their classification, the technology of preparing patterns from plastic materials, their advantages, practical conditions for their production and their basic characteristics are written up and discussed on the basis of foreign literature sources.—M. T.

Works Standards as the Basis of Standardisation in Patternmaking. A. Lenke. (*Giessereitechn.*, 1958, **4**, 135–141). With the object of encouraging increased standardization in patternmaking, a number of illustrative examples are discussed.

Interrelations between Construction, Pattern Design and Foundry. H. Lucas. (*Giessereitechn.*, 1958, **4**, June, 128–132). The necessity for co-operation between constructor, pattern designer and the foundry technical expert is pointed out, with reference to a number of examples.—L. D. H.

Transport of Feed Metal during Solidification of Tapered Steel Bars. E. J. Sullivan, jun., C. M. Adams and H. F. Taylor. (*Trans. A.F.S.*, 1957, **65**, 394–401). Theoretical and experimental studies on the casting of test bars are reported. Convex parabolic distribution of metal is more efficient than straight or concave. Carbon content was varied and sound bars were obtained with steel containing 0·10 to 1·4%C and the amount of taper was also varied.

Feeding Heads on Steel Castings. P. Németh, I. Lipovetz and F. Varga. (*Koh. Lapok Öntöde*, 1958, **9**, May-June, 124–127). In this first part of the paper the authors describe the use and construction of various feeding heads on steel castings.—P. K.

Gating and Risering Shell-Mould Pattern Equipment. D. C. Kidney. (*Mod. Castings*, 1958, **34**, July, 38–40). A practical survey of points to be observed in designing a gating system for shell-mould patterns.

Design and System of Feeding. "A Foundryman." (*Iron Steel Rev.*, 1958, **1**, Feb., 99–101). The design of feeding systems to ensure directional solidification is briefly discussed.—B. C. W.

An Attempt to Establish Norms for Gating Systems, in Order to Enable Them to be Moulded as Refractory Products. T. Sala. (*Prz. Odlew.*, 1958, **8**, (7), 186–194). The possibility is discussed of adopting the system worked out at the Novo-Krematorsk Metallurgical Institute for calculating the dimensions of gating systems for steel castings. The method of calculation of typical casting system dimensions is outlined, and examples are given of tables of standard systems prepared on this basis. The method has practical application in any foundry where there is adequate turnover of sufficiently large castings.

Risers, their Zones of Influence, Permissible Spacing and Mode of Operation. R. Namur. (*Giess. Tech. Wiss.*, 1958, Apr 1077–1087). A general formula for the zone of influence of risers is established theoretically. Calculations are made of the solidification time for any selected position. Examples are worked out for flat plates and rods. The spacing of risers as a function of the size of the casting is considered. Finally, a rapid method of making riser calculations is given.—R. J. W.

Core-Making Practice. M. N. Sarkar. (*Iron Steel Rev.*, 1958, **1**, Feb., 95–98). The manufacture of cores by conventional methods, the shell moulding process, and the CO_2 process are described.—B. C. W.

The Fragility of Chemically Hardened Moulds and Cores. L. Petržela. (*Slévárenství*, 1958, **6**, (5), 142–149). Chemically hardened moulds and cores are generally more difficult to break when freeing the casting than those made with mixtures containing organic binders. It is shown that this deficiency can be largely overcome by adding to chemically hardenable mixtures coal dust, naphthalene and refractory powders in various proportions. Details of the use of such additives and their effects are given.—P. F.

Design Strength of Snap-Flask Moulds and Cores Made from Chemically Hardened Mixtures. M. Bednářík. (*Slévárenství*, 1958, **6**, (5), 149–152). A series of tests on the bend-strength of snap-flask moulds made by pressing, carried out by the author, is described. The results should facilitate a more economical design of CO_2-hardened false cores and moulds.

Thermal Decomposition and Gas Evolution of Phenol Resin for Shell Moulds. S. Minowa. (*Rep. Gov. Ind. Res. Inst.*, 1957, **6**, Dec. 715–718). Pressure, volume and rate of evolution of gases was measured in a vacuum apparatus. The effect of air was also estimated. Optimum curing time was found to be 5 min. at 200–300° C. The rate of gas evolution is decreased by curing and a tar-like carbon is formed at high temperatures in a vacuum. It is considered that addition of a carbonate with a greater decomposition pressure is desirable.

On Measuring Methods in the Shell Moulding Process. V. Collapsibility at Elevated Temperatures. K. Hashimoto. (*Rep. Gov. Ind. Res. Inst.*, 1957, **6**, June, 291–296). [In Japanese]. The time required for a test-piece to collapse under a compressive load at high temperature is measured. For iron alloys a temperature above 800° C is recommended. The collapse is caused by decomposition of the resin and appears to be related to air permeability and flowability of the mixture. Equations relating collapsibility to tensile, compressive, bend and hardness properties are given.

Production of Refractory Moulds by the Shaw Process. (*Technique Moderne*, 1958, **50**, Sept., 404; from *Technische Rundschau*, 1956, Dec. 7). A brief description of this precision moulding method.

The Properties of Phenolic Resins and Phenolic Resin-Sand Systems, Used in Shell-Moulding Practice. G. Ambrus, G. Hevenesi and J. Szekeres. (*Koh. Lapok*, 1958, Jan., Öntöde, 1–5. The properties of phenolic resins and their examination are described, and the effect of these properties, as well as of the amount of resins, sand characteristics and baking on the strength of the shell mould are discussed. For economical reasons the use of a mixture of 40% phenol and 60% cresol resin is suggested.—P. K.

Experience with the Waterglass CO_2 Processes in the Foundry. O. Rácz. (*Koh. Lapok*, 1958, Jan. Öntöde, 5–11). The optimum characteristics of waterglass for the waterglass-CO_2 core-making process, as well as for waterglass-CO_2 moulding are given. For using these methods a minimum temperature of 15° C is necessary in the mould shop.

Recent Developments in the Ferrous Foundry Industry. H. C. Wanner. (*Iron Steel Rev.*, 1958, **1**, Feb., 43–47). Short descriptions are given of the investment casting process, shell moulding and the CO_2 process.—B. C. W.

Chilled Cast Iron. G. Schwarz and D. Junghans. (*Giessereitechn.*, 1958, **4**, Sept., 225–233). An outline of the techniques of

chill-casting, including the types of moulds used and their applications is given, in relation to work carried out in Eastern Germany.

Automatic Moulding Plant for Furnace Castings. W. Gesell. (*Giesserei*, 1958, **45**, Oct. 9, 642–647). An automatic plant is described, which is in full operation, and possesses some novel features. The transport and control systems are described.

Precision Casting by the Lost Wax Process. H. Bode. (*Giessereitechn.*, 1958, **4**, July, 197–200). Various aspects of the manufacture of steel castings by the lost wax process are discussed.—L. D. H.

Ladle Heating in the Foundry. R. B. Renda and W. M. Zeunik. (*Mod. Castings*, 1958, **34**, July, 75–79). Experiments carried out on ladle heating showed that the method normally used in the plant—ladle right side up and covered, with the flame applied at the ladle lip—was most efficient. Of the two types of burners used, i.e. premixing and aspirator, the latter was more efficient.

Centrifugal Casting of Cast Iron Pipes in the Chinese People's Republic. L. S. Konstantinov, B. D. Khakhalin and A. N. Smolyakov. (*Metallurg.*, 1958 (9), 38–39). The article describes the production of 8 in. to 24 in. centrifugally cast pipes at the An'shan plant in China.—R. S.

HEAT-TREATMENT AND HEAT-TREATMENT FURNACES

Change in Dimensions of Bearing Races in Heat Treatment. A. P. Gulyaev and B. M. Zel'bet. (*Metallov. Obrabotka Met.*, 1957, (9), 28–36). A detailed study is reported of the influence of heat treatment and some other factors on distortion in the heat treatment of ball races made from hot rolled annealed tubes, bars and forgings. Ovality was found to increase in proportion to the decrease in the value of width/diameter ratio and increase in hardening temperature from 820–880° C, or oil temperature from 20 to 80° C. Increase in one of the dimensions of the race (except thickness) leads to its proportional increase in heat treatment. Changes in the dimensions of the rates practically do not depend much on hardening or oil temperatures or on previous heat treatment.—S. K.

The Three Key Figures in Quality Constructional Steels. R. Calvo Rodes. (*Inst. Hierro Acero*, 1958, **11**, 9 7–103). [In Spanish.] The author discusses the heat treatment and characteristics of these steels and suggests that there are three key factors that control real equivalence between the steels. Thus, in order to substitute one steel for another, both must (*a*) quench to produce homogeneous martensite in equal thicknesses between equal limits, (*b*) have equal susceptibility to quench cracking under the same conditions of treatment, and (*c*) temper in the ductile range to give similar structures and mechanical properties.—P. S.

Waste-Heat Recovery from a Pusher-Type Furnace Fired with Blast-Furnace Gas. H. Weineck. (**Stahl Eisen*, 1958, **78**, Sept. 4, 1246–1251). Waste heat from a pusher-type furnace fired with blast-furnace gas was recovered by means of economizers from the gas at a temperature of less than 300° C for heating purposes. The equipment and its economics are fully described.—T. G.

Hardening or Normalisation (Thermal Treatment of Dowel Pins for Turbines and Steam Pipes Working at High Pressures). B. Z. Tseitlin. (*Metallov. Obra. Met.*, 1957, (3), 45–50). The Cr–Mo–V steel EI 10 is widely used for the manufacture of fastening accessories for turbines and steam pipes working at temperatures of 500–510° under high pressures. Duration tests under actual conditions of service have shown that these accessories, after normalization with subsequent tempering, have an appreciably higher resistance to relaxation than the accessories hardened and tempered at equal temperatures (improvement treatment).—L. B.

The Development of the Theory of Heat-Treatment Due to the Contributions of Soviet Scientists. V. D. Sadovskii. (*Metallov. Obra. Met.*, 1957, (12), 2–14). An historical review starting with D. K. Chernov's contribution on hardening, published in 1855, which mentions the work of many Soviet scientists in the field of metal physics and follows this up with a brief consideration of various heat treatments based on the theories which have found acceptance.—L. H.

Mechanical Properties of Steel Subjected to Rapid Case-Hardening by High-Frequency Induction Heating. K. Z. Shepelyakovsky, A. D. Assonov and P. A. Lankin. (*Metall. Obrabotka Met.*, 1957, (2), 46–48). High-frequency induction case-hardening can cause overheating when carried out in the range 1050–1100° C. The effect of high case-hardening temperatures on the mechanical properties of some CrMn and CrNi steels was investigated.—R. S.

Dispensing with Primary Heat Treatment of Case-Hardening Steels. I. V. Samolov. (*Metallov. Obrabotka Met.*, 1957, (8), 46–50). The results are presented of an investigation of the influence of preliminary heat treatment on the mechanical properties of the most widely used steels for case hardening. With Type 12KhN3A steel forgings, all the specified properties remained satisfactory independent of the nature of the heat treatment. This was confirmed by further tests with this and types 12Kh2N4A and 18KhNVA steels which showed no change in final properties whatever the previous microstructure. This is taken to show that normalization of case-hardening steels as received is not justifiable at the present level of technology.—S. K.

The Effect of Titanium on the Case Properties of Case-Hardening Steels. J. Prohászka. (*Koh. Lapok*, 1958, **13**, April, 180–186). Hungarian case-hardening experiments with steels with C 0·1–0·18%, Mn 0·35–0·65% and Ti 0·1–0·6% have shown an optimum effect of 0·1–0·2%Ti on the case properties, especially by increasing temperature and simultaneously decreasing time of carburization.—P. K.

Automatic Controls Guarantee Top-Quality Carburizing. P. M. Unterweiser. (*Iron Age*, 1958, **182**, July 3, 61–63). The new, fully automated carburizing and heat treating line of the Timken Roller Bearing Co. is described. There are 2 carburizing furnaces 67 ft long and 12 ft wide. Automatic control of the dew point is the vital feature of these furnaces, and the instruments are described.

Effect of Barium Carbonate in Carburisers on the Quality of the Carburised Case. A. M. Tarasov and M. P. Semenchenko. (*Metall. Obrabotka Met.*, 1957, (5), 49–53). The effect of barium carbonate as a catalyst on the nature of the saturation of steel with carbon during case-hardening is discussed. It is found that, when selecting a solid carburizer, the susceptibility of the steel to carbon saturation must be taken into account. The barium carbonate content of the case carburizing mixture must be kept below 3%.—R. S.

Surface Hardening of Carbon Steels by Gas Carburization. O. K. Kotov. (*Metallov. Obra. Met.*, 1957, (10), 28–33). A detailed investigation of the mechanical properties of three ordinary-quality carbon steels after gas carburization is reported. Spindle oil was found to give a eutectoidal carbon content in the surface layer combined with good post-carburization mechanical properties. On raising the tempering temperature of the carburized steels to 400° C the bending and tensile strengths increase in spite of a considerable hardness decrease. Bending strength decreased when the depth of the carburized layer exceeded 0·4 mm (test-piece diameter 10 mm). The endurance limit of the carburized steels was found greatly to exceed that of the normalized steels.—S. K.

Electric Tempering of Surface-Hardened Parts by Successive Heating. Yu. M. Bogatyrev and S. M. Gamazkov. (*Metallov. Obrabotka Met.*, 1957, (9), 51–58). The conditions for tempering surface hardened parts by induction heating with a moving inductor have been studied in relation to the hardness, structure and residual stresses of the tempered test pieces. The hardness was found to decrease less than with ordinary tempering at the same temperature, and in low-temperature tempering the compressive strains in the hardened layer and the tension in the transition layer decrease to a greater extent. The industrial application of electric tempering is advocated with the use of thermocouples or sensitive paints for the essential temperature measurement.—S. K.

High-Frequency Heating Rates and Quenching Temperatures for Pearlitic Grey Cast Iron. P. I. Rusin. (*Metallov. Obra. Met.*, 1958, (3), 53–56). An iron with 3·45%C, 0·65%Mn, 1·79%Si, 0·14%P and 0·116%S was used with cylindrical specimens 25 mm × 40 mm high. These were heated to 800–1150° C at 100–400° C/sec and depth of hardening was found to depend upon temperature and rate of heating. Quenching from 900–950° C gave different hardness patterns from those obtained by quenching from 1000–1050° C. The metallographic constituents of the hardened zones are given. Optimum conditions and choice of temperature and quenching for various properties are worked out. At the optimum the hard zone was 1·5–2 mm thick and the transition zone 2·5–3 mm.

Surface Hardening of High Strength Cast Irons by High-Frequency Induction Heating. M. N. Kunyavskii, T. G. Demidova and E. N. Nikolaev. (*Metall. Obrabotka Met.*, 1957, (3), 41–45). High-frequency surface hardening of pearlitic cast iron with a nodular graphite structure produces 57 to 58 Rockwell C hardness. Ferritic and ferritic–pearlitic irons with a nodular graphite structure must be subjected to a special heat treatment ensuring saturation of the γ-phase with carbon from the graphite inclusions only in the surface layer. The suggested method of saturating the surface layer of the γ-phase with carbon from the graphite inclusions by means of single or multiple normalizing using high frequency heating with subsequent quenching ensures high surface hardness with a ductile core.—R. S.

Kinetics of the Solution of Tungsten Steel Carbides During Induction Heating. I. N. Kidin. (*Metall. Obrabotka Met.*, 1957, (2), 18–23). The temperature at which Fe_3W_3C dissolves is lower with induction heating than with conventional heating. WC dissolves almost completely in 0·6%C steel at 880° C if the rate of heating is 40° C/sec. When this rate is raised to 75° C/sec., the carbides dissolve completely at 960° C in 1%W steel and at 1040° C in 5%W steel. At rapid heating rates (200° C/sec), the effect of this temperature increase is inversely proportional to the W content of the steel (for a 1%C steel). Carbide particle size is reduced at 980 to 1040° C as the W content diminishes.—R. S.

Report of an Investigation of Hardening Rail Ends by High Frequency Current. Yao Yi-Shun, Kon Sun-Lyn, Shun Io, Chin I-Sun. (*Acta Metallurgica Sinica*, 1957, **2**, (1), 107–126). The new technique developed abroad of hardening rail ends has been tested at the Anshan Combine. It consists of hardening by high-frequency currents, and results show that the physical and mechanical properties of rails so hardened is increased 1·5 times to double, in comparison with non-hardened rails. The changes in structure were also studied.—L. H.

Heating by Electro-Magnetic Induction. R. Anthoine. (*Rev. M. Tijds.*, 1958, **2**, 50–58). Conventional and induction methods of heating are compared with reference to the thermal treatment of metals. Methods of generating high-frequency currents are discussed from the theoretical and practical points of view. The influence of frequency, thermal effect, depth of penetration and power variables are dealt with.—R. P.

Nitridable Titanium Steels. L. Gillemot and M. Tömöry. (*Per. Polytechn.*, 1958, **2**, (1), 1–17). The usual thickness layer aimed for in practice can be obtained quickly with Ti steels. When nitriding, a Ti/C ratio < 4 gives a comparatively soft case; above this ratio the layer is harder, the hardness increasing as the ratio becomes higher.

The Influence of Gas and Salt Bath Nitriding on the Properties of Structural Steels (Part III). H. Wiegand and M. Koch. (*Metalloberfläche*, 1958, **12**, Aug., 225–230). Formerly the applications of gas (ammonia) and salt bath (cyanide) nitriding were in separate fields. Recent investigations have shown that the latter method can be used with advantage for structural steels of suitable composition.

Nitriding of Steel by Heating with High Frequency Current. E. M. Morozova and F. P. Florensova. (*Stanki i Instr.*, 1958, (6), 28–31). It has been established that heating by high-frequency current during nitriding ensures an intensive saturation with N_2 in a much shorter time than when heating in a furnace. A thickness of a nitrided layer of the order of 0·2 mm is reached in about 3 h, whereas it would require 20–24 h in a furnace. The optimum temperature lies between 500 and 550° and it gives a layer of hardness HV 1000–1100. An increase in temperature reduces the hardness.—L. H.

Harder Stainless Practical. R. N. Libsch. (*Steel*, 1958, **142**, June 30, 85–86). A new method of hardening stainless steels is described; it involves activating the surface with a salt bath, after which gas nitriding will produce a hard case to the desired depth. The salt bath treatment involves 15 min. immersion in fused NaCN–KCN. The hardened surfaces retain desirable stainless properties for many applications.—D. L. C. P.

New Facts on the Nitriding of 4140 and 4340. A. J. Schwarzkopf. (*Iron Age*, 1958, **181**, May 29, 90–93). The results of experiments on the effects of nitriding these 2 high strength steels are presented.—D. L. C. P.

Reducing Brittleness of the Nitrided Layer of Russian 38 CrMoAl Steel. A. A. Yurgenson and T. M. Pogrebetskaya. (*Metall. Obrabotka Met.*, 1957, (4), 41–44). Preliminary heat treatment has a marked effect on the brittle strength of the nitrided layer. Water quenching of the CrMoAl steel from 930° C improved mechanical properties more than oil quenching and reduced the brittleness of the nitrided layer. The marked drop in brittle strength of the nitrided layer of specimens quenched from above 1000° C is due to grain growth and the formation of a nitride network.—R. S.

High-Temperature Carbonitriding of Plain Carbon Steels. O. K. Kotov. (*Metallov. Obrabotka Met.*, 1958, (5), 16–20). Studies on two steels with 0·2% and 0·35%C respectively are described, variously treated. Tensile, bend and impact tests were carried out. High-temperature gas carbonitriding was found to be the best treatment, and this improves mechanical properties, especially endurance limit. Wear resistance improved and was not reduced by tempering up to 300° C. Stress distributions are also plotted.

Industrial Application of Gas Cyaniding. B. A. Stetsenko. (*Metallov. Obrabotka Met.*, 1957, (10), 43–48). The author indicates that gas cyaniding is not widely used in Soviet industry. He shows the advantages of gas over liquid cyaniding with examples of its successful use at the Gor'kovskii car works. He considers gas cyaniding practice and conditions for studying the kinetics of the process.

Low Temperature Cyaniding of Dies for Casting under Pressure. S. B. Svishchova and M. S. Barilo. (*Vestnik Mashin*, 1958, (8), 41–42). A die, made of steel 3Kh2V8, after forging, annealing and rough machining, is hardened and tempered, giving a hardness of 28–32 R_C. A process is described for low-temperature cyaniding of the die giving a layer 0·08–0·15 mm thick with a surface hardness of 59–64 R_C ensuring a good strength for the die.—L. H.

Problems of Decarburisation of Electrotechnic Steel Sheets. J. Groyecki. (*Hutnik*, 1957, **24**, Kuly-Aug., 289–295). [In Polish] Various methods of decarburization of electrotechnic steel sheets are surveyed. Special attention is paid to the method of decarburization in an atmosphere of CO–CO_2, H_2–CH_4 and by using scale. The last method applied at 800° C is the most efficient with low carbon steel. Results of decarburization of low carbon steel and silicon steel are given.

The Annealing of Tool Steel. (*Metalen*, 1958, **13**, Apr., 142–147; May, 158–163). [In Dutch]. These sections deal with isothermal transformation, transformation with continuous cooling, annealing tests on a steel with the composition C 0·9%, Mn 1·2%, Cr 0·5, W 0·5%, V 0·2%; the annealing of Cr–Ni–Mo tool steel containing 4%Ni, of the type C 0·45%, Cr 1·5%, Ni 4%, Mo 0·4%.

Effect of Recirculation of Protective Atmospheres and of Annealing Time in Pot Furnaces on the Properties of Cold Rolled Strip Steel, Including the Economy of the Process. J. G. Brockhaus, R. Horst, and A. Lüth. (*Stahl Eisen*, 1958, **78**, Oct. 16, 1449–1456). Recirculation of the protective atmosphere results in reduction of gas consumption, and annealing and cooling time and in an improvement of the mechanical properties of the strip as well as in increased through-put of the plant by reduced heating-up and cooling times in addition to shorter annealing time. Forced recirculation brought about a high uniformity of temperature in the furnace.—T. G.

Fast Annealing of Sheet-Strip Coils with Helium Injection. J. D. Keller. (*Iron Steel Eng.*, 1958, **35**, April, 109–113). The paper describes a method of annealing, which obviates the necessity for circulating fans by increasing the radial conductivity of the coils at least five-fold. This is achieved by injecting He into the narrow spaces between the convolutions of the coils.—M. D. J. B.

Dimensional Changes due to Annealing of Plastically Cold-Worked Metals. VII. On the Effects of Cementite Precipitation of Low Carbon Steel Bar. H. Sekiguchi and M. Inagaki. (*Nippon Kinzoku*, 1956, **20**, Mar., 150–154). A relationship is established for the dimensional change due to annealing of a 0·10%C steel bar cold-worked immediately after water-quenching from 650° C, and the sum of the changes due to annealing of the specimen water-quenched from 650° C and cold-working after pre-annealing at 950° C. The pattern of tendencies is also followed by the hardness and microstructure.—K. E. J.

Open Coil Anneal System Designed for Steel Plant Operation. (*Iron Steel Eng.*, 1958, **35**, June, 151–158). The Lee Wilson opened coil recuperative annealing system is described. It is claimed that heating speeds are faster and temperature uniformity improved. Both technical and operating data are given.

Effect of Isothermal Treatment on the Endurance and Wear Resistance of Steel. M. B. Shapiro. (*Metall. Obrabotka Met.*, 1957, (3), 57–60). Isothermal treatment in the lower part of the intermediate range increases the fatigue strength and the wear resistance of steel. Isothermal treatment was used to increase the life of valve slides in compressors working under conditions of cyclic stressing and wear. Further research is required to solve the question as to why isothermal quenching sometimes increases the impact strength and notch-toughness, but does not increase the fatigue strength although a drop in fatigue strength as a result of isothermal quenching has not been noted.—R. S.

Electrolytic Heating. (XII). Application to Heat Treatment of Machine Parts: I. H. Mii, Y. Sato and S. Funahashi. (*Rep. Gov. Ind. Res. Inst. Nagoya*, 1958, **7**, Oct., 730–742). Treatment of a bicycle crankshaft is described. Press-made C steel shafts were heated and quenched under various conditions and hardness measured and structures observed. Recommendations for practical application are made.

The Annealing of Malleable Cast Iron in a Lorry Factory [FSC] in Lublin. J. Raczka. (*Prz. Odlew.*, 1958, **8**, Sept., 257–262). [In Polish]. Different types of annealing furnaces for white and black-heart malleable cast iron found in Poland are described. A detailed description of an electric chamber furnace, recently installed at the FSC, is given of capacity 26·3 t and max. power 650 KW an analysis of the annealing process, the chemical composition of the blackheart malleable iron used and its mechanical properties after annealing are given. A detailed analysis of operational costs is made in comparison with analogous costs in different types of annealing furnaces in Poland. The hope is expressed that the time of annealing and operational costs might be considerably cut in the near future.

Graphitisation in Salt Baths; its Application to High-Sulphur Cast Irons. O. Bader and D. Gadot. (*Fond. Belge.*, 1958, April, 123–125; *Fonderie*, 1958, **148**, May, 211–214). The authors state that graphitization of white iron increases with temperature. They advocate the use of barium chloride baths with additions of cyanide, carbonate, chlorides or borax at 1000°–1050° C for graphitizing high-sulphur iron. Salt baths prevent oxidation and excessive warping.—R. F.

Water-Soluble Quenching Oils. K. H. Kopietz and J. G. J. van Overvest. (*Metalen*, 1958, **13**, Aug., 273–279). The properties are described of some new quenching oils which can be rinsed off the work with water, with which they form an emulsion. Trials are described investigating the quenching action of these oils, and the effect on quenching of the presence of small amounts of water. Results were found to be satisfactory.

Investigations of the Cooling Properties of Quenching Oils. S. Orzechowski and H. Chelminska. (*Prace Inst. Hutn.*, 1958, **10**, (1), 17–27). Seven types of mineral oil and three methods of investigation were used. The method of Grossmann and Asimov gave data not in agreement with that of Grossmann and interpretation by Russell's method showed the coefficients found to depend on the test-piece diameter. The oils were classified in three groups, but the methods of measurement are not regarded as satisfactory, and further work is necessary.

State of Progress of the IRSID Work on Transformation Curves in Isothermal and Continuous Quench Conditions of Steels Made in France. A. Constant. (*Centre Doc. Sider. Circ.*, 1958 **15**, (5), 1085–1094). A description is given of the aim of the investigation and of some of the results obtained to date. The practical application of the results obtained under continuous quench conditions are discussed. A further 30 curves under isothermal and 31 curves under continuous quench conditions will be published shortly.—B. G. B.

Isothermal Quenching of 0·5% Carbon, 1% Chromium Spring Steel. I. V. Trush. (*Metall. Obrabotka Met.*, 1957, (3), 60–62). Isothermal quenching of spring steels reduces warping. Experiments on 0·5%C, 1·0%Cr steels subjected to isothermal quenching without tempering are reported. Results of hardness and fatigue tests are given. The spring steel tested exhibited a lower fatigue strength when subjected to isothermal quenching without tempering than when it was oil-quenched followed by tempering. Holding in the isothermal bath for more than ten min did not increase the endurance limit of the steel.—R. S.

Interrupted Quenching of Carbon Steels. B. I. Shubov. (*Metall. Obrabotka Met.*, 1957, (3), 55–57). Isothermal or interrupted quenching is not suitable for carbon steel parts. Experience has shown, however, that small parts up to $1\frac{1}{2}$ in. diameter can be treated with a fairly slow transformation of the austenite to the required structure. This can be attained by means of interupted quenching using severe quenching media such as spray or salt solutions. Interrupted quenching in the temperature range corresponding to the transformation of austenite to acicular troostite produces the best combination of mechanical properties.—R. S.

High-Power Quenching Media. L. V. Petrash. (*Metalloved. Obra. Met.*, 1958, (3), 56–61). Water and water-base agents were used and errors in the Rose curves were found. Quenching in hot water was found to be irregular and to be more effective at higher temperatures (120–220°). Additions of salt and alkali shorten the film vapourization stage. Concentrated NaOH quenches more slowly than water below 300° C and NaCl solutions below 230° C. Heating the solutions has not the irregular effect of heated water. Cold 10–15% solutions of NaCl and

15–30% NaOH are recommended for parts liable to cause local superheating. Breakdown of the vapour film is due to surface cooling below the critical point and is assisted by oxide or salt films and by moving the parts in the quenchant and by lowering bath temperature. The incidence of quench cracking is discussed.

Studies on Quenching Media. IX. On the Severity of Quench Determined by Using Cylindrical Steel Specimens. M. Tagaya and I. Tamura. (*Nippon Kinzoku*, 1956, **20**, June, 336–339). Results for the severity of quench using various media with carbon tool steel and ball-bearing steel, obtained by Grossman's method, are reported. (13 references). **XI. Correlation between Cooling Curves for a Silver Specimen and the Quenching of Steels.** (Aug., 428–432). The quenching severity of any medium towards steel may be rapidly assessed by reference to a correlation chart; this is drawn up using silver specimens which are easily measured. (15 references). **XII. On the Severity of Quench, *H*, of Various Quenching Media. XIII. Reconsideration of the Cooling Ability of Mineral Oils.** (Sept., 477–480; 480–484). XII. For use in determining hardenability, values of the severity of quench (*H*), are determined from cooling curves with silver specimens for twelve animal and vegetable oils. These are compared with tricresyl phosphate and dioctyl phthalate. (11 references). XIII. The *H* values of mineral oils increase with temp. to a max. and then decrease; these max. values are correlated with flash point. Paraffinic mineral oils have smaller *H* values than naphthenic oils.—K. E. J.

Study of Quenching Strains Set Up in High-Chromium Steels. C. Rovirosa. (*Mét. Constr. Mécan.*, 1958, **90**, June, 473–481). This section deals with the influence of the origin of the steel and its thermo-mechanical history on the quenching strains, and studies in detail the quenching technique. Seven steels were investigated, all with a similar composition: 1·5 to 2·1%C, and 11 to 13·3%Cr.

Effect of Secondary Quenching from the Intercritical Range on Proneness to Reversible Temper Brittleness. B. G. Sazonov. (*Metall. Obrabotka Met.*, 1957, (4), 30–34). It is shown that structural steel becomes unstable as regards reversible temper brittleness after preliminary normal quenching and secondary hardening from the intercritical range.

The Tempering of 2¼%Cr–1%Mo Steel after Quenching and Normalizing. R. G. Baker and J. Nutting. (*JISI*, 1959, **192**, July, 257–268). [This issue].

ROLLING MILL PRACTICE

On the Formation of Longitudinal Cracks in the Foot of Rails on Straightening. S. Hořejš. (**Hutn. Listy*, 1958, **13**, (6), 517–526). On the basis of detailed researches described in the paper, it is concluded that the cracks occur as a result of the application of excessive stresses on straightening, but only if they are nucleated by rolled-in sub-surface blow-holes. Their propagation is also assisted by unfavourable local rolling textures. Even the avoidance of excess force in straightening was found to lead to the almost complete prevention of crack-formation of this type.—P. F.

Departures in Metal Rolling. A. I. Tselikov. (*Engineering*, 1958, **186**, Oct. 17, 521–522; from *Izvestia Vyshikh Uchebnykh Zavedenii. M.V.P. USSR, Mashinostroenie*, 1958, (1)). Both automation and the substitution of rolling for other forming methods are discussed. "Screw" rolling and rolling under tension are mentioned.

Double Cooling Bed for Rod Mill. D. H. Driscoll and C. Renberg. (*Iron Steel Eng.*, 1958, **35**, March, 141–143). The article describes a mechanized bar cooling bank specially designed to give efficient cooling and increased output of a rod mill.—M. D. J. B.

Turn-Up and Turn-Down in Hot Rolling. G. E. Kennedy and F. Slamar. (*Iron Steel Eng.*, 1958, **35**, March, 71–76). The paper describes an investigation that was made to determine whether the curvature of a bar emerging from a mill could be controlled during rolling. The authors claim that the most significant factor affecting turn-up and turn-down is the surface-speed ratio of the top and bottom rolls.—M. D. J. B.

On the Tendency to Tear Formation in Rolling Semi-Ferritic Chromium Steels. M. Zezulová and L. Koutník. (*Hutn. Listy*, 1958, **13**, (5), 412–418). Plant and laboratory experiments designed to elucidate the cause of tear-formation on rolling Cr steels containing 0·15%C max., 0·90%Mn max., 0·70%Si max., and 15·5–18·5%Cr are discussed. It is concluded that the tendency to tear formation is influenced by the nature of the two-phase structure of the steel, i.e. by the quantity and distribution of the alpha phase and carbides. At the rolling temperature regions of this phase mixture transform into the gamma phase, the distribution of these regions depending upon the original distribution of the carbides. The greater the non-homogeneity of the steel as to distribution and size of carbide agglomerates, the greater is also the non-homogeneity of the alpha-plus-gamma structure at the rolling temperature and the tendency to tear formation during rolling. This tendency can be removed or diminished by aiming at an initial structure which should be as homogeneous as possible, particularly with regard to the distribution of carbon and thus of carbides. Further, it is desirable to use a comparatively low temperature for the finishing passes so as to obtain a fine-grained structure and a homogeneous carbide distribution. In removing tears by grinding, the grinding direction is best chosen to be perpendicular to the rolling direction.

Rolling Textures of Transformer Steels. H. Möller and H. Stäblein. (*Arch. Eisenh.*, 1958, **29**, June, 377–390). The textures of cold, hot and cross rolled specimens of transformer steels containing 2·65 and 2·85%Si were investigated. Texture was measured using a specially-developed X-ray Geiger counter "texture goniometer," using cobalt K_α radiation. The influence of various factors on the accuracy of the pole figures is discussed. (36 refs.).

Some Problems Encountered in Improving Hot-Rolled Transformer Sheet. V. Rauner, J. Skála and M. Zezulová. (*Výzkumné Práce z Oboru Železa a Oceli, State Publ. Ho. Tech. Lit., Prague*, 1958. 157–171). [In Czech]. The literature relating to factors affecting the watt-losses of transformer iron is reviewed. The author's own experiments showed that in twice poured melts, whether pre-deoxidized with Al prior to the addition of ferro-silicon or not, watt-losses in a saturating field of 10 kG can be reduced to 0·07 W/kg by the reduction of the carbon content to 0·015%. The reduction can be carried out by heating in clean scale. Embrittlement was also studied. It was found to be associated with large ferrite grains and local peaks in the concentration of silicon and carbon.—P. F.

Rolling and Heat Treatment of Textured Transformer Steel Sheets. A. G. Pietrienko. (*Hutnik*, 1957, **24**, July-Aug., 277–282). [In Polish]. The effect of the contents of Si and C, the size of the grain and other technological factors, on the magnetic and plastic properties of cold-rolled transformer steel sheets, are discussed.

Production of Dynamo and Transformer Sheets. E. Neuhoffer. (*Kohászati Lapok*, 1958, **13**, Jan. 6–10). The production technique of dynamo and transformer sheets in Hungary is described. For hot-rolled transformer sheets the author suggests the following chemical composition: C 0·05 max., Mn 0·10 max., Si 4·20–4·60, P 0·020 max., S 0·005 max., Cu + Ni 0·10% max. Al up to 0·3% is permissible, even desirable, for improving ageing resistance. For cold sheets the optimum Si content is 3%. In order to avoid complaints, examination and grading of every sheet is recommended.—P. K.

Improvement of the Quality of Hot-Rolled Transformer Sheets. V. Rauner, J. Skala and M. Zezulova. (*Vyzkumne Prace z Oboru Železa a Oceli*, 1958, 157–171). Melting processes were studied and the possibility of reducing carbon content during annealing is discussed; the content can be lowered to < 0·02% and possibly further during rolling.

Development of Automatic Thickness Controls for Strip Mills. I. G. Orellana. (*Iron Steel Eng.*, 1958, **35**, March, 144–150). The author reviews current systems in use for controlling strip thickness in cold mills. Some of the difficulties in controlling thickness on a hot strip mill are discussed.—M. D. J. B.

Wallingford Steel Co. Adds New Mill Equipment. (*Iron Steel Eng.*, 1958, **35**, March, 176–178). The article describes the installation of a second Sendzimir mill to roll precision strip in widths up to 27 in. and as thin as 0·001 in. for special applications such as turbine engines and guided missiles.—M. D. J. B.

Destruction of the Concrete Foundation of a Rolling Mill by Oil Emulsion. H. Kaminski. (*V.D.I.-Z.*, 1958, **100**, July 11, 851–854). The concrete foundations of a rolling mill may be heavily attacked by lubricating oil or aqueous emulsions, in combination with cooling water and the high temperatures from the mill. Protective measures are suggested; these should be incorporated in the design of new plant.

Cold Rolling of Tin-Plate Sheet at the Basse-Indre Forges. P. Gombaud. (*Technique Moderne*, 1958, June, 291–296). The author describes the machinery and method used for producing cold-rolled steel strip for galvanizing, starting from hot-rolled material. The main stages in this process are: pickling, rolling, degreasing and heat-treatment. The starting material is 680–910 mm thick, coils weighing about 4·5-t. The final product going to the galvanizing plant is about 2 mm thick and the coils weigh about 9 t.—R. P.

Influence of Cut Edges on the Properties of Cold-Rolled Strip. P. S. Sakharov. (*Metalloe. Obrabotka Met.*, 1957, (8), 43–46). An account is given of investigations of the effects of burred cut edges, of their filling to various extents and of heat treatment on the tensile and cold-brittleness properties of steel strip. Most tests were on 3 × 10 mm strip of type 1Kh25Yu5 steel (0·1%C, 25·1%Cr, 5·6%Al). A depth of 0·1 mm had to be removed to restore tensile properties to their pre-cutting level. Rounding of the edges largely restored the upper temperature of the cold brittleness range. Heating to 650° C followed by water cooling restored plastic properties, but this raised the cold brittleness temperature limit. Similar results were obtained with strip of other types of Cr–Al steels with a tendency to cold brittleness and those with their cold-brittleness limit near room temperature. The behaviour of some other steels is briefly considered.—S. K.

Production of Weldless Steel Tubes by the Stiefel Process. J. Verdickt. (*Technique Moderne*, 1958, **50**, Sept., 395–401). The processes of preparing the bars, piercing, rolling, finishing and checking are outlined, with diagrams.

WELDING AND FLAME-CUTTING

Experimental Welding of Railway Wagons in Low Alloy Steels. I. I. Zhdan. (*Svarochnoe Proizvodstvo*, 1958, (11), 35). Experimental construction of railway wagons by welding in low alloy carbon steels is discussed. Satisfactory welds were produced at a high level of output.—M. T.

The Effect of Vibration of the Electrode on the Process of Arc Welding and the Properties of the Welds. A. A. Alov and V. S. Vinogradov. (*Svarochnoe Proizvodstvo*, 1958, **31**, (9), 19–22). Experiments were made on low-carbon steels with vibration at frequencies up to 60 cycles/sec and at varying amplitudes, and it was found that this had a marked effect on the welding process and the mechanical properties of the welds. Owing to the transference of the metal in small droplets, arc stability was increased, burning through of the metal was decreased, the configuration of the welded joints was improved, and the welding of parts of small thickness was facilitated. The structure of the metal of the welds formed with vibration of the electrode had a

much finer grain and was more even. The mechanical properties of the welds were improved. Their susceptibility to the formation of porosity, cracking and similar defects was reduced. Vibration has a favourable effect on all fusion methods of welding. It has an especially marked effect during argon-arc welding owing to the absence of other means of metallurgical influence on the weld metal, during CO_2 arc welding owing to the more complete slag formation and removal of the oxides forming, and during electric slag welding by the reduction of the quantity of dendrites.—M. T.

The Effect of Niobium on the Structure of Welded Joints in 18–8 Austenitic Steel. N. F. Lashko. (*Avtom. Svarka*, 1958, (6), 84–87). Niobium is used to prevent intergranular corrosion and increase the refractoriness of such steels, but it may give rise to high-temperature weld cracking. Its action is directly connected with its distribution between the solid solution and the primary and secondary precipitation phases. Lashko, contary to other authors, maintains that in low-carbon 18–8 austenitic steels to which Nb is added with a normal content of alloying elements, there can only be formation of the following phases: γ and α solid solutions, metastable ε-phase, the carbide phase (Cr, Fe)$_{23}$C$_6$ (if the Nb content is small), the carbide (Nb, Cr)C or carbo-nitride phase Nb (C, N), the σ-phase and also the intermetallic phase Fe_2Nb, similar in structure to Zn_2Mg. It is maintained that the suggested H-phase of Mukhum and Pal'chuk is not isomorphic with the χ-phase ($Fe_{85}Cr_{12}Mo_{10}$), but a crystalline structure of the type Fe_3W_3C, and is in reality an oxide of the type (Cr, Fe, Ni)$_3$Nb$_3$O with a parameter of 11·25 Å.—M. T.

The Formation of a New Phase in Welded Joints in 18–8 Steel containing Niobium. G. G. Mukhum and N. Yu. Pal'chuk. (*Avtom. Svarka*, 1958, (6), 88–91). The authors maintain that the H-phase has a parameter of 8·84–8·94 kX, but admit that it and the χ-phase are not isomorphic. The H-phase is formed in welded joints which have been stabilized at 850° or after prolonged heating at 650° C. It is considered that it is equivalent to M_6C, and cannot be an oxide as suggested by Lashko. The formation of this phase, which occurs at Nb contents above 2%, explains the fall in ductility and impact strength of welds with Nb contents over 2% after tempering at 1100° and 1000 h heating at 650° C; under the same conditions the σ-phase is not formed. Final confirmation is awaited in the shape of X-ray analysis of an individual crystal of the phase.—M. T.

The Welding of Stainless Steel. B. Mulder. (*Lastechniek*, 1958, **24**, June, 95–100). Various technical aspects of the welding of stainless steel workpieces are discussed.

Influence of Titanium and Nitrogen on the Structure and Properties of a Welded Joint on Thin High-Chromium Steel Sheets. D. A. Odesskii and V. M. Vozdvizhenskii. (*Metallov. Obrabotka Met.*, 1957, (9), 42–46). The micro-structure and mechanical and service properties of welded joints made by different methods in high-Cr. steels alloyed with Ti and N have been investigated. The steels, Kh28T and Kh28N, had the respective compositions: 0·09, 0·013%C; 25·05, 26·0%Cr; 0·30, 0%Ti; 0·46, 0·45%Si; 0·54, 0·74%Mn; 0·012, 0·012%S; 0·027, 0·023%P; 0, 1·61%Ni; 0, 0·20%N. Considerable growth of ferrite grains in the seam area occurred in Kh28T, excess phases separating round the grains. This gave poor joints. Better joints were obtained with Kh28N, where grain growth was slower and limited by an envelope of austenite: this steel has good corrosion-resistance properties.—S. K.

Welding of Alloy Steels without Preheating in the Chemical Industry. W. Gilde. (*Schweisstechnik*, 1958, **12**, June, 84–87). To avoid difficulties due to preheating in the welding of high carbon and alloy steels the method due to Dechner and Speicher was developed, in which delayed cooling is employed. This procedure is said to simplify welding and to save time. Several examples are discussed.

Fabrication Welding of Heavy Section Vessels by the Electro-Slag Process. W. Anders. (*Schweisstechnik*, 1958, **12**, June, 65–70). The technique of the fabrication of thick-walled pressure vessels (plate thickness > 1½ in.) using the electroslag welding process is described. Longitudinal seams are necessary to operate the process; where circumferential seams are required CO_2 shielded arc welding is used.

Dynamic Behaviour of Direct-Current Apparatus for Arc Welding. Recommendation for Assessing the Applicability to Welding. M. A. Carrer. (*Rev. Soudure*, 1958, **14**, (3), 146–154). The relation between requirements to be met in arc welding and the electrical apparatus used is discussed in detail. Variables due to the electrical circuit are examined. Recommendations made are based on results of oscillographic data recorded during welding.—R. P.

The Use of Low-Alloy Grain-Refined Steels in Welded Structures. E. M. Kuzmak and V. S. Milanchev. (*Svarochnoe Proizvodstvo*, 1958, **31**, (9), 11–15). A study was made of the thermal reactions of the welding process on the mechanical properties of such steels, and it was found that such properties and those of the welded joints are dependent on the conditions of their heat treatment. A method is outlined of calculating the appropriate welding parameters based on the theory of heat distribution during welding, which is capable of ensuring an even strength of the welded joint. Thermal grain refinement of low-alloy steels for welded structures enables them to be used to greater advantage.—M. T.

A Method of Assessment of the Resistance of Steels to the Formation of Low-Temperature Cracking after Welding. N. N. Prokhorov and E. L. Makarov. (*Svarochnoe Proizvodstvo*, 1958, **31**, (9), 15–18). Experiments on corner welds and testing techniques are described, which were carried out at up to 500° C and down to reduced temperatures on fractures with a brittle fine-grain structure in several steels of varying compositions. Further investigation was also made of methods of avoiding such fractures, e.g. by heat treatment of the electrodes. The method of assessment is shown to give sufficiently accurate and consistent results, and may be used for correlating data by its practical application.—M. T.

On Changes in the Composition of the Slag during the Electric Slag Welding Process. D. A. Dudko and I. N. Rublevskii. (*Avtom. Svarka*, 1958, (6), 51–55). During this welding process definite changes occur in the migration of additions along the line of the weld, and the absolute value of this migration is dependent on the material used (flux and electrode wire), the nature and polarity of the welding current, and the welding parameters which determine the character of the transfer of the electrode material in droplets. Research into compositional changes of the slag provides a means of forming a concept of the course of the metallurgical reactions which occur during the welding process. Oxides of iron are accumulated in the slag. Mn undergoes reaction in the molten pool practically without loss through vaporization, but appreciable quantities of Si are always lost in this way in the form of volatile fluorides. Likewise the concentration of CaF_2 in the slag is reduced and the CaO content is increased as a result of the migration of fluorine into the gaseous phase.—M. T.

Special Features of the CO_2 Arc Welding Process using Thin Electrode Wire. I. I. Zaruba and A. G. Potap'evskii. (*Avtom. Svarka*, 1958, (6), 32–41). Consideration is given to the demands which the process makes on the dynamic characteristics of supply sources during welding, and recommendations are given for the choice of welding generators. At the parameters normally employed, this welding method is accompanied by a large number of short circuits of the arc gap (100–150/sec.). The number of short circuits and the size of the electrode metal droplets are above all dependent on arc pressure (arc length). Instability is evidenced by variations in the arc gap, the arc pressure and the welding current, and a certain periodicity of the change in these factors. Reliability of arc excitation and stability of the welding process are basically affected by the rate of current rise after short circuits of the arc gap and its fixed value; the optimum values of these factors are given. The relationship between the nature of the transfer of the metal and the size and rate of the short circuit current rise is indicated.—M. T.

The Influence of Low-Frequency Vibration on the Crystallisation of the Metal in the Molten Pool and the Properties of the Weld Metal. V. L. Russo and P. N. Efimov. (*Svarochnoe Proizvodstvo*, 1958, (11), 10–12). Experiments were carried out on steel and Al alloy sheet at frequencies of 25, 46 and 55 Hertz and at 10 amplitudes between 0·01 and 2·3 mm. Amongst the basic factors governing the effect of low-frequency vibration on crystallization of the molten pool are the specific pressure impulses set up during crystallization, at a certain value of which the ends of the growing dendrites are destroyed; these crystal fragmentations form additional coercive crystallization centres, and result in refinement and disorientation of the primary crystallization structure. In the range of frequencies and amplitudes investigated, an increase in frequency led to a reduction in primary grain size and an increase in impact strength. Welding of Al alloys in this range can be carried out under normal welding parameters. For steel amplitudes of 0·5 to 1·5 mm can be used to change the primary crystallization structure. To obtain still greater grain refinement experiments should be made with the use of vibration at supersonic frequencies.—M. T.

How Can the Formation of Martensite be Prevented in Arc Welding? (*Tekn. Ukeblad.*, 1958, **105**, June 12, 566–568). Trials are described in which the possibility is investigated of preventing the formation of martensite in arc welding by controlling the rate of cooling with an auxiliary burner. The method can be applied to automatic welding.

Advances in Welding Technique in the Construction of Rolling Mill Plant. O. Wilmes and J. Brück. (*Schweissen Schneiden*, 1958, **10**, June, 253–255). Examples are given of the increasing tendency for the use of welded frameworks for rolling mill plant.

Research into the Process of Pressure Welding of Tubes. V. D. Taran, N. V. Bobritsky, A. S. Fal'kevich and I. E. Neifel'd. (*Svarochnoe Proizvodstvo*, 1958, (11), 12–14). In order to determine whether pressure welding is brought about by diffusion or by the formation of metallic bonds between atoms on the surfaces to be welded with diffusion playing only an auxiliary role, a study of the process was made with the use of radioactive isotopes. Micro-radiography was carried out on welded joints in low-carbon Mn-steel tubes 30 mm in dia. and wall thickness 4–7 mm, by fusion and induction heating with high-frequency currents. No appreciable diffusion through the former interface was observed. No grain growth through the interface occurs, welding has a non-diffusional character, and fusion takes place through the proximity of the surface atoms at distances which promote the formation of bonds between the atoms of metal. Heating to a high temperature and upsetting promote this proximity. Owing to the predominately non-diffusional character of the process, very little advantage is gained from post-weld heat treatment, apart from slight grain refinement and removal of the effects of superheating. Diffusion does of course occur to a limited extent, but its role is purely secondary.

High Penetration Electrodes. C. P. Felgueroso. (*Cien. Tecn. Solda.*, 1958, **8**, May-June, title 3, pp. 14). The early and post-war development of deep-penetration electrodes is outlined, theoretical aspects are discussed and practical aspects described and illustrated. Some examples of fabrication using this technique are given.—L. D. H.

Testing the Effectiveness of Fluxes. II. V. Ruža. (*Zváranie*, 1958, **7**, (6), 181–183). [In Slovak]. The efficiency of several

Czechoslovak and foreign brazing fluxes was studied in standard tests, and the most suitable and economic fluxes for various applications were thus determined.—P. F.

Use of the Transformation Diagram for Questions of Choice of Materials for Welding and Flame Hardening. F. Wever, A. Rose and L. Rademacher. (*Forsch. Wirts. Nordrhein-Westfalen*, 1955, No. 170, pp. 51). The use of the TTT-diagram is illustrated.

Autogenous Strain at Low Temperatures. K. Krekeler, H. Verhoeven and H. Ernenputsch. (*Forsch. Wirts. Nordrhein-Westfalen*, 1958, No. 531, pp. 36). Measurements of strain removal under welding conditions are reported.

Qualitative Investigations of Welding by Means of the Automatic Arc Using Bare Wire and the Use of Ferromagnetic Powder as Sheathing. K. Krekeler and H. Verhoeven. (*Forsch. Wirts. Nordrhein-Westfalen*, No. 274, 1956, pp. 55). The method is outlined and is shown to present no difficulties. Sheet 15 mm thick can be welded without special precautions, but thicker sheets take longer. X-ray and other tests were made including mechanical tests.

Sigma-Phase in 25/20 CrNi Steel Weld. B. I. Medovar. (*Metall. Obrabotka Met.*, 1957, (3), 27–30). The best way to increase ductile properties of 25/20 CrNi steel welds operating at 700–875° C is to raise the operational temperature up to 1000° C–1100° C for a short time about every 500 to 600 h. This dissolves the σ-phase and ductile properties which have been temporarily lost are recovered.—R. S.

On the Structure and Properties of the Heat-Affected Zone of Welded Joints in 17Cr2Ni Steel. N. I. Kakhowski. (*Avtom. Svarka*, 1958, (6), 64–68). In view of the expense, laboriousness and difficulty of exercising control of preliminary heat treatment of parts to be welded, it was decided to experiment with arc welding of this steel (up to 8 mm thick) without such treatment, which has been previously recommended by H. Thilsch, *Welding J.*, 1955, No. 1), and also to investigate the optimum post-weld heat treatment, during which particular attention was given to the study of the impact strength and structure of the heat-affected zone at various heat-treatment temperatures. 17Cr2 Ni steel up to 8 mm thick was satisfactorily welded without preliminary heat treatment. In the heat-affected zone an inhomogeneous structure is formed. Alongside soft grains of low-carbon ferrite there are hard grains of acicular ferrite, constituting metal of extreme brittleness. To soften this metal of the heat-affected zone, and in order to obtain welded joints with satisfactory properties, post-weld, high-temperature heat heatment at 700 ± 20° C is essential.—M. T.

The Heat-Resistance of Austenitic Welded Joints Obtained by Submerged Arc Welding, and Argon and CO_2 Arc Welding. B. I. Medovar, A. N. Safonnikov and R. O. Lents. (*Avtom. Svarka*, 1958, (6), 13–31). When welding 18–8 and 25–20 Cr–Ni austenitic steels, although one and the same basic materials (steel, electrode wire) are used, the long-term refractoriness of the weld metal may vary in respect of the welding method used. Welding of 1Kh18N9T steel, under a shield of fluoride oxygen-free flux ensures a very much higher long-term strength than under a shield of low-silicon manganese flux or argon arc welding. The refractoriness of the weld metal was higher when obtained by CO_2 arc welding owing to a certain degree of carburization of the weld metal (by 0·02–0·03%) and its hardening by carbide formations, though in the absence of powerful carbide-forming elements (Ti, Ni, Mo) in the weld, this tendency to increase the refractoriness is only slight. In submerged arc welding it is also possible to obtain an increase in refractoriness, by increasing the carbon content of the weld metal (up to 0·12–0·15%). Argon-arc welding of 25–20 steel produced better refractoriness than welding under a shield of fluoride flux. No direct connection was discovered between the oxygen content, non-metallic inclusions in the welded joint, composition, form and nature of the distribution of these elements in the weld, and refractoriness of the joints.—M. T.

MACHINING AND MACHINABILITY

An Electrolytic Jet Machining Technique for the Production of Thin Foils of Steel. P. M. Kelly and J. Nutting. (*JISI*, 1959, **192**, July, 246–248). [This issue].

PROTECTIVE COATINGS

Adherence of Zinc on Heavily Galvanized. Hard Drawn Wire of Rimming, Low-Carbon Steel. D. Horstmann. (*Stahl Eisen*, 1958, **78**, Oct. 16, 1456–1462). Adherence increases with the presence of N and P in the steel. The usual accompanying elements in this grade of steel do not exercise an effect on adherence, but slight, uniform roughening of the surface increases it. Adherence decreases with increasing zinc bath temperature and galvanizing time. The test procedure and the results obtained are reported in detail.

Sherardizing. (*Draht*, 1958, **9**, April, 132–136). The process is described and its protective quality is discussed. There follows a comparison of sherardized and electro-galvanized coatings on bolts, nails, and the like.

Performance of Various Chromate Supplementary Films on Electrodeposited Zinc Coatings. H. L. Katz, K. L. Proctor and F. Nagley. (*Proc. A.S.T.M.*, 1957, **57**, 203–220). Accelerated test methods were studied and correlated with service performance in marine environments. Ten commercial processes were tried and depletion rate of hexavalent Cr and the contribution of tervalent Cr were followed.

Residual Stress in Steel Strip for Tinplate. Y. Kuwahara and R. Iwamoto. (*Toyo Kohan*, 1957, **5**, Dec., 23–30). [In Japanese with summary and captions in English]. Measurements by the electrolytic stripping method are reported. Effects of heat-treatment were traced. The sources of residual stress are shown in a diagram.

Modern Developments of Hot-Tinning Techniques in Industry. W. E. Hoare. (*Met. Corr. Ind.*, 1958, **33**, Apr., 184–188).

Hot Dip Tinning Operations at Port Kembla. J. Guthrie. (*Proc. Australasian I.M.M.*, 1958, Sept., 27–54). An illustrated account of the plant and process.

Electrolytic Deposition from Electrolyte Melts of Bright Tin Precipitates on Sheet Iron. A. I. Vitkin and T. P. Plotnikova. (*Doklady A. N.*, 1958, **120**, (3), 588–591). Experiments in tin plating in relation to time interval, temperature and concentration of tin chloride, have suggested that a bright surface of the tin layer may be obtained without additions of substances, but only through the application of an external current.—S. I. T.

Ultrasonic Plating. Y. Miyamoto. (*Toyo Kohan*, 1957, **5**, Dec., 37–42). [In Japanese]. The use of ultrasonics in electrolytic tin plating on steel strip in a stannous sulphate bath was studied. A smoother and brighter finish was obtained and higher current density could be used and thicker coatings obtained. The electron microscope did not, however, reveal any differences. Corrosion (by the thiocyanate method) was more pronounced with the ultrasonics-treated sheet, and this appears to be related to the occurrence of wavy banding on these sheets.

A Study of the Thiocyanate Porosity Test. T. Ando and H. Morimoto. (*Toyo Kohan*, 1957, **5**, Dec., 31–36). [In Japanese]. Electro-tin plates were examined. Acid concentration, pre-cleaning and temperature and time of testing were studied. The tin–iron electrode potential was measured in acid media.

Study on Aluminium-Coated Steel. II. On the Formability of Aluminium-Coated Steel Sheet. T. Saga. (*Nippon Kinzoku*, 1956, **20**, Mar., 121–124). For good forming properties in the sheet, the thickness of the layer of Fe–Al compound should be approx. 5–10μ, or up to 15μ if the hardness can be reduced. This is best done by adding over 3%Si to the Al bath as inhibitor. **III. On Cracks in the Alloy Layer and Oxidation at High Temperature.** T. Saga and O. Miyakawa. (Apr., 184–188). Cracking in the alloy layer is greater with rapid cooling rates. It is also greater the thinner the alloy layer, although it decreases slightly with thicknesses of < 50μ. The thicker the Al layer, the less cracking. Cracking adversely affects the heat-resistance.—K. E. J.

Antimony Plating on Steel and Zinc. G. R. Schaer, W. H. Safranek and C. L. Faust. (*Plating*, 1958, **45**, Feb., 139–143). Acid pre-treatment reduces porosity and improves adhesion, especially treatment with H_3PO_4. The plated steel is more resistant than Zn-coated steel to salt spray fog. Outdoor exposure tests are inconclusive.

Protection at a Distance Given by Flame Sprayed Metallic Coatings on Steel. H. Spindler. (*Werks. Corr.*, 1958, **9**, May, 278–283). Experiments were carried out by flame spraying steel surfaces with coatings of Zn, pure Al and an Al–Mg alloy, leaving gaps in the protective coatings. In some cases the coatings were covered with paraffin. Corrosion resistance was tested by total and by intermittent immersion in sea-water, and stress corrosion tests were also carried out. A distance effect was observed, related in general to the ratio of covered to exposed area.—L. D. H.

Flame-Spraying Plastisols. R. F. McTague. (*Plating*, 1958, **45**, Mar., 245–246). Advantages of flame-spraying are noted as heavy coatings can be applied without tendency to sag. However equipment is bulky, the open flame may be hazardous in some surroundings and at low temperatures surface pre-heating is almost impossible. Tests on a submarine are described.

Mechanism of Passivation and Surface Layer Formation. T. Heumann. (*Arch. Eisenh.*, 1958, **29**, Aug., 505–512). The passive behaviour of metallic substances was considered in relation to the passivation processes on pure iron, chromium and other metals. In the passive state corrosion takes place with iron at a rate independent of potential. Thickness of the passive layer increases with increasing potential, corrosion taking place by diffusion of the iron ions through the almost non-porous passive layer. The passivation region of chromium steels in sulphuric acid is greatly increased by small additions of Ni, Mo and Cu.

Investigations of the Formation of Surface Layers in the Cathodic Polarisation of Steel Surfaces. H. Klas. (*Arch. Eisenh.*, 1958, **29**, May, 321–328). Current density-time curves were plotted at constant polarization potentials, using sections of steel tubing immersed in electrolytes containing Ca and Mg ions, and simulating diluted and undiluted sea and drinking waters. The results are explained on the basis of the formation of secondary layers in the cathodically polarized steel surfaces. Conclusions are drawn regarding suitable measures for cathodic protection of the steel.—L. D. H.

Some Current Applications of Phosphate Coatings. E. P. Koeckritz, jun. (*Plating*, 1958, **45**, March, 248–251). Various uses are mentioned in which they replaced Cd, Ni etc. The materials mentioned are mostly non-ferrous.

Modern Methods of Surface Protection. P. Knežević. (*Zaštita Mat.*, 1958, **6**, (3), 121–126). [In Serbocroat]. Principles, applications and advantages arising from phosphatizing are discussed. The "Hydro-Finish" cleaning process utilizing a water-borne suspension of abrasive is considered.

Organic Finishes over Metal Surfaces. M. Perez. (*Plating*, 1958, **45**, Mar., 239–244). An account of the use of synthetics in which recommendations for the applications of epoxides, phenolics, acrylics and vinyl dispersions are made.

New Metallizing Processes. P. Colomb. (*Zaštita Mat.*, 1958, **6**, (3), 103–106). [In Serbocroat]. The utilization of powdered metal in organic resinous binders for use in protective surface coatings is discussed.—P. F.

Porcelain Enamel. A.S.T.M. Committee C-22. (*A.S.T.M. preprint*, 1958, (33), pp. 4). A revised glossary is given.

High Quality Enamel Ware. (*Swedish Foreign Commerce*, 1958, (5), 37–39). An account of the products of Kockums of Ronneby, founded, 1840.

Practice of Applying Paints. (*Usine Nouv.*, 1958, **14**, May 22, 95–97). A brief outline of the method of preparing the work and applying the paint in the treatment of metal structures.

PROPERTIES AND TESTS

Research at the Liége Section, Centre National de Recherches Métallurgiques. O. L. Bihet and P. Coheur. (*Booklet commemorating 10th anniversary, supplement for visiting scientists* 1958, pp. 52). **Laboratory Researches.** (9–24). Physico-chemical analysis, spectrography, analysis of gases and inclusions; physical metallurgy, optical and electron microscopy, ageing of low-carbon steels, mechanism of resistance to deformation at high temperatures, uses of Co. X-Ray diffraction studies of deformation and recovery in single crystals, precipitation structures in hard metals. Analytical and structural problems. (20 refs.). **Applied Research.** (25–35). Development of metal constructions by relieving the working stress of carbon steels A37 and A42. Weldability of steels especially for ship construction. Exposure tests on mild steels. **Works Researches.** (37–47). Conversion processes, refining of MM and OM cast irons, dephosphorization in the ladle, deoxidation with various ferro-alloys and Ca–Si, hydrogen in basic Bessemer steel, blast furnace studies (17 refs.). **International Iron and Steel Industry Research.** (49–51).

The Significance of Test Results from Small Groups of Specimens. E. H. Schuette. (*Proc. A.S.T.M.*, 1957, **57**, 1304–1320). An account of a statistical method with tables of "exceedance" and directions for plotting and interpolation.

The Use of Statistical Analysis in Metallurgy. M. Knotek, J. Likeš, and R. Vojta. (*Vyzkumne Prace z Oboru Železa a Oceli*, 1958, 7–26). [In Czech with summaries in English, German, and Russian]. Methods of arranging data and calculating the normal distribution curve are shown and the design of experiments is considered. Examples are then given of the interpretation of the mechanical properties of steels and of the use of point analysis for determining the relative quantity of a phase in unit volume. Other applications are mentioned.

The Mechanical Properties of Hardened High-speed Tool Steels. A Review of Methods of Determination. J. Clayton-Cave and E. Ineson. (*JISI*, 1959, **192**, July, 248–253). [This issue].

Metal Surface Finish. L. Walter. (*Met. Fin. J.*, 1955, **1**, Apr., 168–176). A review of methods of measurement.

A New Method of Determining the Microgeometry of a Surface. A. I. Kvacheva and N. D. Druchak. (*Vestnik Mashin.*, 1958, (3), 54–55). A method based on direct electron projection is described, which renders it possible to measure the accuracy of the surfaces not only of flat but also of curved specimens.—L. B.

Some Metallurgical Properties of Mild Steels and Their Influence on Surface Finish. A. W. F. Comley. (*Met. Fin. J.*, 1955, **1**, July, 311–317, 326). Various grades of steel are defined and the effects of surface finish, ageing, defects from presswork, and pick-up are indicated. (12 refs.).

The High Purification of Aluminium, Uranium, and Iron by the "Melted Zone" Technique. P. Albert. (*Met. Ital.*, 1958, **50**, April, 136–142). [In Italian]. The author describes the melted zone method of purifying Al, U, and iron and reviews the new properties exhibited by these metals in their state of extreme purity. The work described was largely carried out at the Vitry Laboratory of the French National Centre for Scientific Research. (19 references).—M. D. J. B.

The Electrochemical Properties of the Stressed Layer of the Surface of Pure Iron. N. Ohtani. (*Nippon Kinzoku*, 1956, **20**, July, 400–403). Results are given for variations in thickness, electrode potential and free energy of activation of the hydrogen electrode reaction for stressed layers produced by polishing, filing, sawing, turning etc.—K. E. J.

The Structural Theory of the Flow of Metals. A. Oding. (*Rev. Mét.*, 1958, **55**, May, 448–452). According to the theory here developed, creep rate is dependent upon the number of dislocations ready to begin movement. On this basis a general equation is developed for the creep curve which is in agreement with experimental data.—A. G.

The Importance of the Elastic Properties of Spheroidal Graphite Cast Iron for the Construction of Machines. C. Gianola. (*Giesserei*, 1958, **45**, Apr. 24, 210–219). The mechanical properties of spheroidal graphite cast iron, such as impact toughness, elongation, and reduction-in-area are discussed. The elastic properties, namely, the modulus of elasticity and the limit of proportionality are then described, together with the hardness of spheroidal graphite cast iron. The carrying out of experimental tests is described and the results tabulated. Examples of the application of this material are given and illustrated in eleven photographs.—R. J. W.

Investigation of the Relation of the Modulus of Elasticity of Iron to Temperature. R. I. Garber and A. I. Kovalev. (*Zavods. Lab.*, 1958, (4), 477–479). The dynamical method of investigation permits the measuring of the modulus of elasticity in deformations of the order of 10^{-5} — 10^{-3}. The quantitative characteristic of plasticity in this method is Poisson's ratio, which increases suddenly at 750° and reaches its maximum at 900°.—L. B.

Deformation of Solids at High Rates of Strain. A. H. Cottrell. (*Chartered Mech. Eng.*, 1957, Nov., 448–460). A lecture. Dislocations are dealt with most extensively.

Studies on Silico-Manganese Spring Steels. Part IV: Effect of Carbon Content on Flexure Properties. S. Jayaram, H. R. S. Rao, V. G. Paranjpe, and S. Visvanathan. (*Tisco*, 1958, **5**, April, 41–45). The authors determined the influence of carbon content on the mechanical behaviour of oil-quenched and tempered Si–Mn steels from the results of flexure tests. It was found that the optimum properties corresponding to the conditions of stress implied by the camber test could be obtained only if the minimum carbon content was fixed at 0·55%. This value coincided with the limit determined on the basis of end-quench hardenability tests. The authors recommend that the carbon range of oil-hardening Si–Mn spring steels should be raised to 0·55–0·65%. **Part V: Temperability.** S. Jayaram, H. R. S. Rao, and V. G. Paranjpe. (46–52). The authors investigated the tempering characteristics of Si–Mn spring steels. It was concluded that (1) the temper hardness of such steels containing 0·41–0·77% C in the range 350° to 680° C, tempering time 1 h, can be expressed by a simple equation, (2) the effect of time on temper hardness follows the logarithmic rule, (3) a wide range of values for the constant c in the Holloman-Jaffe tempering parameter can be chosen without affecting the accuracy of (2); (4) one value of c can be used for the entire range of carbon contents, and (5) the contribution of alloying elements towards temper hardness is lower at higher levels of carbon.—L. E. W.

Damping Measurements on Weakly Stretched Iron–Carbon Alloys. F. Wever and W. Wepner. (*Forsch. Wirts. Nordrhein-Westfalen*, 1957, No. 457, pp. 22). Carbide precipitation from iron with < 0·02% C after straining is examined.

Observations on the Dislocation Structure of Sub-boundaries in Iron. F. W. C. Boswell. (*Proc. A.S.T.M.*, 1957, **57**, 521–526). The development of etch pits on dislocations in vacuum-melted iron is described and electron micrographs of replicas are shown.

Coulomb Excitation of Iron-57. G. F. Pieper and N. P. Heydenburg. (*Phys. Rev.*, 1957, **107**, Sept. 1, 1300–1302).

Excited States in ^{59}Co and ^{56}Fe. M. Mazari, A. Sperduto, and W. W. Buechner. (*Phys. Rev.*, 1957, **107**, July 15, 365–367).

Excitation Function for ^{56}Fe$(n, p)^{56}$Mn. J. Terrell and D. M. Holm. (*Phys. Rev.*, 1958, **109**, Mar. 15, 2031–2035).

Precision Determination of Some Energy Levels in ^{56}Fe, ^{65}Zn, and ^{99}Tc. E. L. Chupp, J. W. M. DuMond, F. J. Gordon, R. C. Jopson, and H. Mark. (*Phys. Rev.*, 1958, **109**, Mar. 15, 2036–2040).

Superexchange Interaction Energy for Fe^{3+}–O^{2-}–Fe^{3+} Linkages. M. A. Gilleo. (*Phys. Rev.*, 1958, **109**, Feb. 1 777–781).

Proposed Procedure for Testing Shear Strength of Brazed Joints. F. M. Miller and R. L. Peaslee. (*Weld. J.*, 1958, **37**, Apr., 144s–150s). The problem of determining the strength of a brazed joint under shear is discussed and it is concluded that the published data on the shear strength of brazed joints is misleading and of little value to design engineers. A proposal is advanced for a standard specimen and some results obtained using this are presented. Compression shear tests are briefly considered and the relation between the results obtained with the proposed test bar and the design of welded joints is discussed.—R. C. W.

Theoretical Aspects of the Plasticity of Metals. A. Vach. (*Hutnik*, 1958, **8**, (6), 203–206). The concept of dislocations, yield criteria of polycrystalline aggregates, the resolution of stresses into hydrostatic and shear components, and other theoretical aspects of the plasticity of metals are discussed, and their bearing upon such practical problems as the calculation of deformation in forging, and the effect of temperature changes on the plasticity of metals are considered in an elementary manner.—F. F.

Accuracy of the Erichsen Cupping Test. V. Kokkonen and G. Nygren. (*Jernkont. Ann.*, 1958, **142**, (10), 638–659). Tests were made at several stations, in the second series the tools and the functioning of the machines were checked and the tests supervised to ensure agreement with the DIN 50101 conditions. Significant differences were found between machines in both series of tests.

Residual Stresses in Surface-Hardened Parts After Tempering at Temperatures Corresponding to the Break-Down of Martensite. M. M. Kobrin, and K. V. Shishokina. (*Metall. Obrabotka Met.*, 1957, (5), 29–33). Examination of a steel containing 0·93%C, 1·4%Cr, 0·12%Ni, 0·35%Si, 0·33%Mn, 0·024%S and 0·026%P showed a change in the residual stresses from compression to tension in the hardened case after tempering at temperatures corresponding to the intensive break-down of martensite. There is also a change in the residual stresses of the transition zone beneath the hardened case. Tensile residual stresses are transformed to compression stresses. The redistribution of residual stresses with a change from tension to compression is of a local character embracing the hardened case and the transition zone.—R. S.

The Successes of Soviet Metal Physics. A. A. Astaf'ev. (*Metallov. Obra. Met.*, 1957, (11), 2–17). On the occasion of the 40th anniversary of the revolution this publication gives a review of the achievements of Russian scientists in this field and enumerates the many types of steel which have been evolved and developed giving the names of the scientists and investigators who have contributed to this development.—L. B.

Analysis of Residual Stress in Ground Surfaces of High-Temperature Alloys. R. D. Halverstadt. (*Trans. A.S.M.E.*, 1958, **80**, May, 929–940). Distortion and reduced endurance limits arising from grinding have proved troublesome. A complete analysis of wheel speed, hardness, fluid, feed rate and work speed effects were made for 3 gas-turbine alloys. Recommendation for stress reduction are made.

Stress Concentrations Around a Small Spherically Isotropic Spherical Inclusion on the Axis of An Isotropic Circular Cylinder in Torsion. S. B. Dutt. (*J. Techn.*, 1958, **3**, June, 13–17).

On the Molten Metallic State. E. Scheil. (*Giess. Tech. Wiss.*, 1958, Jan., 1007–1019). A molten metal is considered as an atomic fluid. Theories of van de Waals, Lennard-Jones and Borelius are given and discussed. Comparison is made between the three aggregate states in the neighbourhood of the melting point. The importance of the minimum in the P–V curve for crystallization effects is stressed. The speed of crystallization, the number of nuclei and the time taken for solidification are discussed. The theories of viscosity and of diffusion are described at length. Finally molten alloys are classified.

Evaluation of a Single-Shear Specimen for Sheet Material. W. W. Breindel, C. L. Scale and R. L. Carlson. (*A.S.T.M. preprint*, 1958, (80), pp. 7). A specimen giving ultimate shear strength values is shown and evaluated for 304 stainless steel and non-ferrous alloys. Design is such that values for tearing or buckling are higher than for shear.

Shape and Dimensions of a Tensile Test-Piece for Sintered Metals. I. Relations between the Thickness of a Test Specimen and the Tensile Strength and Elongation of the Materials. E. Hayashi and H. Horasawa. (*Rep. Gov. Ind. Res. Inst.*, 1957, **6**, Dec., 708–714). Density distribution diagrams were also obtained. Between 3–7 mm tensile strength and elongation decrease with increased thickness. Up to 5 mm this is gradual but over 5 mm it becomes very rapid. The lower the density the greater the effect of thickness.

The Mechanical Properties of Grey Iron in the Light of Statistics. N. Hajto. (*Kohaszati Lapok*, 1958, **13**, Feb.-Mar., Öntöde, 47–52). The mechanical properties of grey iron, such as ultimate tensile strength, bending strength and deflection, Brinell hardness, as well as their correlation to the degree of saturation are investigated on frequency distribution curves, plotted from about 2000 test results.—F. K.

The Problem of the Strength Characteristics of Test Bars Cast-On and Cut from the Casting, as Applied to the Examples of Cast Steel. H. Resow and W. Trommer. (*Giesserei*, 1958, **45**, July 3, 387–397). After discussing the factors influencing the values obtained for the tensile strength of steel castings, the results of trials carried out on test bars cast on and cut from the casting were compared and were found to be in good agreement. Hence a cast-on test bar is adequate, and it is unnecessary to cut specimens from the casting.

Investigations of the Deformation Process in Iron–Silicon Alloys in its Realtion to Composition and Temperature. F. Lihl and E. Buhl. (*Archiv. Eisenh.*, 1958, **29**, June, 365–375). The temperature relationship of the mechanical properties of Fe–Si alloys containing 1·05 to 6·10%Si was examined in relation to static tensile and impact bending tests, deformation-temperature and impact toughness-curves being derived for various silicon contents.

The Determination of S-Curves by the Method of Continuous Cooling. J. Váňa and D. Tlustá. (*Výzkumné Práce z Oboru Železa a Oceli, State Publ. Ho. Tech. Lit., Prague*, 1958, 119–137). [In Czech]. The TTT-curves of nine medium and low-alloy steels, mainly as used in the construction of vehicles, were determined by dilatometry and by continuous cooling, utilising in the latter the modified end-quench method of C. A. Liedholm. (*Trans.A.S.M.*, 1946, **38**, 180–208). The information derivable from the diagrams is discussed.—F. F.

Mechanical Properties of Cast Titanium–Iron and Titanium–Aluminium–Iron Alloys. N. Hehner, H. W. Antes and R. E. Edelman. (*Trans.A.F.S.*, 1957, **65**, 374–378). As-cast tensile and impact properties are given. Tensile strength increase with Fe content and could be as high as 147 000 with both alloys. Ductility is reduced with increase of Fe as is impact strength. Hardness increases. Rate of hardening of the ternary alloys is affected by Al content.

TTT-Diagram of High Chromium High Carbon Ledeburitic Die Steel. K. Nakazawa. (*Nippon Kinzoku*, 1956, **20**, Apr., 195–198). Data were obtained for steel containing 2·55% C, 11·8%Cr and 0·65%W austenitized at 900°, 970° or 1050° C for 10 min. The pearlite transformation is enhanced by the nucleation of undissolved carbides, but the bainite transformation is not. The M_s points corresponding with the austenitizing temperatures were 210°, 170° and 70° C, respectively.

On the Determination of the Mechanical Properties of Steels for Deep Drawing by the Numerical Values of Power Coefficients. B. S. Natapov and E. S. Fal'kevich. (*Zavods. Lab.*, 1958, (8), 1013–1014). The most important quality characteristics for cold-drawing sheet steels are: yield stress, the ratio yield stress/tensile strength, hardness and the Ericsson tests. The power coefficient was measured with an electrodynamic meter (*ibid.*, 1947, (6), M. A. Grabovskii) and the specific grain size by the method of Saltykov (see *ibid.*, 1946 (7), (8)). The power coefficient method eliminates the errors inherent in the present methods of determining yield stress and tensile strength.—L. H.

Influence of Polarization on the Mechanical Properties of Steel. J. V. Karpenko and R. I. Kripyakevich. (*Doklady A.N.*, 1958, **120** (4), 827–829). Tensile testing was carried out on steel samples immersed in an electrolyte and subjected to cathode and anode polarization, and the results obtained were compared, in each case, with those of testing in air. The variation in the values of mechanical properties of steel was traced to the nature of the electrolyte, the current density, the nature of the anode and other factors.—S. I. T.

Stress–Strain Diagrams for Steel with Direct Load Application. A. B. Litvin. (*Zavods. Lab.*, 1958, (10), 1239–1242). Ordinary tensile tests are performed on ordinary machines at a variable speed of deformation. The author calls such machines class I machines. In real conditions the load increase is often independent of the deformation. To realize conditions similar to real ones in the Nikolaev Shipbuilding Institute a tensile test machine was constructed, which permits the variation of the rate of change of load from 0 to 800 kg/sec; this rate can be kept constant. The maximum load is 3 300 kg. The machine is classified as a class II machine. Some tests were performed with steel specimens; the rate of change of stress varied from 4·25 kg/cm²/sec to 960 kg/cm²/sec, with times of tests from 6 to 21 min. Comparative graphs and results of tests carried out on class I and class II machines are given.—S. S.

Some Factors Influencing the Area-Load Characteristics for Semismooth Contiguous Surfaces Under "Static" Loading. F. F. Ling. (*Trans. A.S.M.E.*, 1958, **80**, July, 1113–1120). The ideal case of "Semismooth" contiguous surfaces is studied theoretically, i.e. one of the surfaces is considered to be rigid and macroscopically smooth while the other is allowed to have a number of wedge-shaped asperities. The area-load characteristics under static loading and their relationship to theories of friction are discussed, and factors causing these characteristics to be non-linear are examined. (25 refs.).

Technological and Mechanical Properties of Various Deoxidised Thomas Steels. R. Rosegger. (*Berg-Hütten Monatsh.*, 1958, **103**, Sept., 163–175). The influence of deoxidation with Si, Si + Al, and Al alone on the degree of cleanliness and the mechanical properties of Thomas steels was investigated. Results of tensile, doubling, notch-impact, weld bending and cupping tests indicated a quality, in the case of low-inclusion Al-killed basic Bessemer steels which is comparable only with fine-grain O.H. steels.

The Effect of Structure on Funnel-shaped Fracture in Drawn Steel Wire. T. Nishioka. (*Nippon Kinzoku*, 1956, **20**, Apr., 181–184). [In Japanese, with summary in German.]

Comparison of Semi-Empirical Solutions for Crack Propagation with Experiments. J. Frisch. (*Trans. A.S.M.E.*, 1958, **80**, May, 921–928). Steels and Al alloys were examined under uniaxial tension. Empirical formulae were found accurate enough for establishing relative merits for materials. Neither energy-release nor stress-concentration criteria gave reliable constants.

Size Effects in Slow Notch-Bend Tests of a Nickel–Molybdenum–Vanadium Steel. J. D. Lubahn and S. Yukawa. (*A.S.T.M. preprint*, 1958, (79), pp. 16). Specimens of various sizes and notch radii cut from a forging gave strength values which were as low as 40 000 psi with large specimens and sharp notches. The notch root strain at fracture decreased with increasing size in geometrically similar specimens. Static cracks caused no more weakness than did sharp machined notches. Effects of notch depth, temperature, hydrogen content and proximity to welds were also shown.

Effect of Size on the Yielding of Mild Steel Beams. C. W. Richards. (*A.S.T.M. preprint*, 1958, (75), pp. 13). The systematic variations of the upper yield point with specimen size are investigated and the theory is extended to pure bending. A substantially linear relationship between the logs of size and yield is demonstrated in agreement with the theory and supports the brittle fracture analogy for the yielding of mild steel.

On the Toughness-Change of Bearing Steel by Heat-Treatment. R. Tanaka. (*Nippon Kinzoku*, 1956, **20**, May, 251–255). Bearing steel SUJ2 was examined by static bending tests. In oil-quenching and marquenching, max. hardness was obtained after austenitizing at 900° C for 5 min., whereas toughness decreases with higher temp. of austenitizing, and above 900° C the superiority of marquenching in increasing toughness diminishes. Austenitizing times should be short. Subzero treatment increases hardness but reduces toughness. Marquenching followed by tempering up to 200° C gives better results than oil-quenching so followed.—K. E. J.

Torsional Fatigue Properties of Small Diameter High Carbon Steel Wire. H. C. Burnett. (*A.S.T.M. preprint*, 1958, (70), pp. 11). Machines for testing short compression springs (8 at a time), for straight wire under constant amplitude of torsional load and under constant amplitude of torsional deflection and 3 steels were tested. No direct correlation with tensile properties was found. Comparison of a commercial cold-drawn wire with one from vacuum-melted steel showed that the former was affected by its inclusions. Oil tempered wire was less susceptible to longitudinal shear than cold-drawn music wire and shot-peening increased fatigue life of springs.

Impact Test and Brittle Fracture. W. Späth. (*Stahl Eisen*, 1958, **78**, Oct. 30, 1615–1619). The author discusses some problems of impact testing with respect to a prediction of the brittle-fracture tendency of steels. The effect of strain-hardening on impact toughness, the occurrence of the Bauschinger phenomenon during impact tests, the correlation of impact toughness with the area of the crystalline spot and the nature of brittle fracture are reviewed. (14 references.)—T. G.

A Method of Determining the Impact Strength and Cold Shortness of Welds Made by Hand and Semi-Automatic Welding. M. M. Kraichik and N. V. Pashkov. (*Svarochnoe Proizvodstvo*, 1958, (11), 7–10). Where a welded joint consists of a single layer only, the impact strength and cold shortness of the deposited metal will determine the strength of the weld, and these are the most important of the mechanical properties. Where the joint consists of several layers, its strength is determined by the properties not only of the metal deposited by the final pass which is not heat affected, but also of the underlying layers which are heat-affected by the subsequent passes. The strength of the weld, however, is still primarily determined by the impact strength and cold shortness of the metal deposited by the last pass. It is therefore essential that tests of mechanical properties should be completed over a cross-section of the weld at right angles and parallel to the axis of the weld. When making tests on the surface layer at right angles to this surface, account must also be taken of the depth of this layer. The authors consider that tests carried out

over a cross-section parallel to the axis of the weld give far more reliable indications of the strength of the weld, since they are not influenced by the depth of the surface layer. The conclusions outlined are based on the practical experiments described.—M. T.

The Influence of Heat Treatment below the A_1 Temperature (and the Degree of Cold Working) on the Properties of Low-Carbon Steel for Boiler Plates. W. Cias (*Prace Inst. Hutn.*, 1958, (2), 95–108). Research into the affect of quenching from a temp. below the A_1 point on the mechanical properties of steels for boiler plates was undertaken to test the expediency of the Polish technical acceptance specification for such plates, which requires a bending test to be carried out. The following were examined: 1. the influence of quenching in water from 650° C on the results of the bending test; 2. the influence of this treatment on impact strength; 3. its influence on tensile strengths; 4. the influence of a heating temperature below the A_1 point, with either quenching or slow cooling, on mechanical properties; 5. the influence of temp. and time of ageing of steel No. 3443 after quenching from 680° C on mechanical properties; 6. as no. 5, but after 11% reduction by cold working; 7. the tendency of the steels investigated to ageing after cold working; 8. the affect of the degree of reduction on the tendency to ageing; and 9. the effect of spontaneous ageing of steel KB–1 after heating at 650° C and quenching on its mechanical properties. It was concluded that bending tests provide no criterion of quality as an acceptance test for boiler plates. In addition, while ageing as a result of cold work produces an increase in hardness and a reduction in ductility which are, in general, permanent and remain unaffected by working conditions, the changed properties resulting from quenching and ageing have a tendency to revert at high temperatures to their initial values before ageing. As a result, cold working of boiler plate may prove dangerous under service conditions.—M. T.

Cold Working Moves into High-Output Uses. D. J. Davis. (*Iron Age*, 1958, **181,** May 29, 85–87). The increasing use of cold working techniques is accounted for and described. The following have allowed wider applications: better lubricants, better knowledge of alloys, improved dies. Cold pressing is used on a large scale to form piston pins and tappets, and cold-rolling to form shafts and splines.

Work Hardening, Recrystallisation and Softening of Alloyed Austenite. L. A. Metashop and M. E. Blanter. (*Metall. Obrabotka Met.*, 1957, (5), 15–25). The processes of precipitation and solution of carbides are independent of the process of softening when previously work-hardened austenite is heated. This is because, in these two processes, particles take part whose mobility varies with temperature. While the process of carbide formation is associated with the diffusion of highly mobile carbon particles, the processes of softening of previously work-hardened austenite are governed by the migration of atoms of the basic lattice. The author then discusses the relationship between the degree of plastic deformation and the mobility of metallic atoms of the basic crystal lattice, and the softening of Mn-containing austenite.

Cold Working and Mechanical Properties of Hardened Steels. (*Metalloberfläche*, 1958, **12,** June, 170–171). Experiments are described to measure hardness losses of cold worked steel after tempering at various temperatures, with different amounts of cold work. Two types of steel were investigated.—L. D. H.

Influence of Cold Deformation. M. Massin. (*Centre Doc. Sider. Circ.*, 1958, **15,** (5), 1095–1106). The influence of cold rolling on the mechanical properties of 3 grades of O.H. steel in the tempered and untempered conditions is described. Cold rolling was found to increase the elastic limit by 10 kgm/mm² in quenched and tempered samples.—B. G. B.

Slow Strain-Hardening of Ingot Iron. A. W. Sleeswyk. (*Acta Met.*, 1958) **6,** Sept., 598–603). Armco iron was strained at a rate of 10^{-4}/sec and anomalous results were obtained which are discussed. The serrated stress-strain curves observed with this type of material are shown to be a consequence of the anomalous velocity dependence on strain-hardening. A modified theory of blue brittleness is proposed. (20 refs.).

Hysteresis and Anelasticity in Cold-Worked Stainless Steel. J. D. Lubahn. (*A.S.T.M. preprint*, 1958, (77), pp. 11). Creep and cyclic loading tests show hysteresis is not an anelasticity phenomenon. Its effect as a source of damping is discussed.

Studies on the Bauschinger Effect of Soft and Medium-Hard Steels. H. Hoff and G. Fischer. (*Stahl Eisen*, 1958, **78,** Sept. 18, 1313–1320). Cold-worked basic Bessemer steels of 50–75 kg/mm² U.T.S. and cold-worked soft basic Bessemer and O.H. steels showed the Bauschinger effect the more strongly the higher was the U.T.S. Annealing below 250° C had no effect on the appearance of the Bauschinger effect, at higher annealing temperatures and corresponding reduction of special stresses the Bauschinger effect could be eliminated. The tests conducted under a variety of conditions are described in detail and results are presented in many graphs.—T. G.

Influence of Mechanical Working on the Surface-Layer Quality of Turbine Blades of Alloy E1617. V. N. Shakhurin. (*Metalloc. Obrabotka Met.*, 1957, (10), 23–27). The influence of machining factors on the microhardness at various depths, degree of work hardening, surface quality and microstructure of a heat-resisting alloy has been studied.—S. K.

The Softening of Plastically Deformed Metals. V. N. Danilov. (*Metall. Obrabotka Met.*, 1957, (6), 15–16). The results of investigations of the softening of plastically deformed low-carbon (0·03%C) steel, commercial Al and Cu show, generally, that the greater the deformation, the more rapidly softening occurs.—R. S.

The Modes of Deformation Leading to Fracture of Metals, A Review of Progress to January 1957. R. Doldon. (*U.K. Atomic Energy Authority, Industrial Group*, 1957, IGR–TN/C–569, pp. 10). Observations on Al are described.

Investigations of the Influence of Temperature and Rate of Deformation on the Resistance to Deformation of Some Steels. V. Valorinta. (*Werkstattstechn. Maschinenbau*, 1958, **48,** Aug., 452–456). Experiments carried out on a plain carbon and some low and high-alloy steels showed that a general relationship could be established between resistance to deformation, rate of deformation and temperature.

First Results of Research on the Hot Working of Steel. Application of Specially Designed Equipment. C. Rossard and P. Blain. (*Rev. Mét.*, 1958, **55,** June, 573–594). Initial results are given of tests under torsion at temperatures ranging from 900 to 1 200° C, using specially designed equipment, which is described. The results are discussed.

Procedure for Determining Graphically Transitory Phenomena Occurring during the Hot Deformation of Steel. C. Rossard and P. Blain. (*Rev. Mét.*, 1958, **55,** June, 595–598). The authors describe a graphical method of obtaining, from a series of stress-deformation curves derived at constant speed, stress variation as a function of deformation for any law of deformation rate.

A Tentative Guide for Fatigue Testing and the Statistical Analysis of Fatigue Data. Committee E–9 on Fatigue. (*A.S.T.M. S.T.P. No.* 91-A, 1958, pp. 80). A supplement to the manual STP No. 91. Test procedures, selection of specimens and mathematical methods for the analysis of data are set out in detail.

The Application of Statistical Methods to the Determination of Fatigue Curves for Steel Wire. B. Baranowski. (*Prace Inst. Hutn.*, (3), 153–160). The opinion is expressed that fatigue curves prepared on the basis of a limited number of experiments are unsatisfactory, and the statistical nature of fatigue is noted. On the basis of investigations of the fatigue strength of 2·5 mm steel wire for mine hoists (U.T.S. 138·5 kg/mm²), statistical assessment of the results by Mood and Dixon's 'staircase method' was carried out. Full details are given of the method of calculation, and a comparison is made between the results obtained and previously published results. It is claimed that the method of statistical assessment of experimental results, although laborious, permits reliable conclusions to be drawn.—M. T.

Slipping-Clutch Fatigue-Testing Machine. N. B. Owen and H. G. Cox. (*Engineering*, 1958, **186,** July 18, 84–85). A 15 000 lb direct-stress machine is described. Besides bending and torsion tests some modes simulating failures in service have been made.

An Example in Collaboration for Fatigue Testing. St. Nadasan. (*Met. Constr. Masina*, 1958, **10,** (7), 599–600). [In Rumanian]. The article gives an example of collaboration on the widest scale in this important field. Under the initiative of the Applied Mechanics Section of the University of Timisoara, a meeting was called in Bucharest of representatives of all technical laboratories, of the metallurgical research institutions, of the State Research Institute for Heavy Industry of the Railway Research Institute and also of representatives of the Bureau of Standards. Work is at present only in its initial stages.—I. H.

A High-Temperature, Vacuum, and Controlled-Environment Fatigue Tester. G. J. Danek jun. and M. R. Achter. (*A.S.T.M. preprint*, 1958, (73), pp. 4). Equipment for reverse bending at elevated temperature in vacuum was constructed and large amplitudes and low frequencies produced failures in $\sim 10^5$ cycles. A magnetic coupling transmitted motion through the vacuum seal using either oscillating magnets or two electromagnets alternately excited. The machine stops automatically when a crack develops. Type 316 stainless was tested at 3×10^{-5} mm and 1500° F.

An Investigation of the Fatigue Characteristics of Leaded Alloy Steels. G. W. Brock and G. M. Sinclair. (*U.S. Dept. Commerce; PB* 121834, 1957, *Mar.*, pp. 55). The effect of Pb on the tensile and fatigue properties of SAE 1018, 1045, 8620 and 4340 steels was examined using statistical methods of evaluation. The effect was small except at u.t.s. above 130 000 psi. Above the endurance limit leaded SAE 1080 and 4140 tended to give longer fatigue lives on smooth specimens.

Fatigue Phenomena during Combined Rolling and Sliding. A. Nikonoff. (*Rev. Mét.*, 1958, **55,** June, 567–572). Equipment is described for reproducing in the laboratory fatigue fractures caused by combined rolling and sliding. The results are discussed, and recommendations made for the choice of a suitable steel quality for railway wheel tyres subject to severe braking.

Distribution of Fatigue Failures in Flat Hardened Steel Test Bars. W. S. Hyler, L. P. Tarasov and R. J. Favor. (*A.S.T.M. reprint*, 1958, (69), pp. 11). After various abrasive surface treatments bars were tested in cantilever bending and found to fracture near the location of maximum stress and to be relatively insensitive to grinding or abrasion, to relative stress level and to nucleation at or below the surface. Inclusions were studied in bars where failure was predominantly subsurface; 93% had inclusions at the origin and $\sim 55\%$ of these were of oxide type, the rest were Ti cyanonitride. Mean diameter was 0·0037 in. and average depth 0·0041 in. for oxide and 0·0006 and 0·0011 for cyanonitride inclusions.

Fatigue Failure. I. I. W.—Commission XIII. (*Rev. Soudure*, 1958, **14,** (3), 155–157). Fatigue failures in welded parts subjected to vibrations are discussed briefly. No deformation is observed near the fatigue crack which is perpendicular to the surface. A 0·5–0·55%C steel is not easily weldeable unless preheated to 300–400° C, followed by stress-relieving after welding. Butt welding is preferred to fillet welding for such steels if they are subjected to vibrations.—R. P.

Studies on Heat-Checking of Iron and Steels. I. Experiments on Heat-Checking of Hot Die Steels. I. T. Fujiwara. (*Nippon Kinzoku*, 1956, **20,** May, 247–251). For steels containing 0·35%C, 5%Cr, 1·5%Mo and 1%Si, disc

specimens were repeatedly heated and cooled by induction hardening and mandrels were used for hot-pressing 0·3%C steel. The formation of heat-checks is affected by intergranular oxidation of the surface in the heat-treated state. The depth and width of cracks follow an exponential distribution. Heat-checks are propagated gradually to the core, mainly about the boundaries of the primary crystals.—K. E. J.

Investigation of the Influence of Surface Defects on the Fatigue Endurance of Crankshafts. (*Verlag Stahleisen M.B.H., Dusseldorf publication*, 1958, pp. 48). Investigations extending over five years were carried out on large crankshafts, thirteen of which were subjected to torsional and three to bending repeated stresses. All of the crankshafts had shown indications of defects under magnetic flux testing, but, in spite of the long duration of testing; in no case were these previously observed defects found to be the origin of subsequent fatigue failure. Full details of the trials and results are given, with numerous illustrations.

Fatigue Strength of Specimens Built Up by Submerged-Arc Welding. L. A. Glikman and A. N. Babaev. (*Metallov. Obrabotka Met.*, 1957, (10), 37–43). The experiments described confirmed the deleterious effect of depositing metal by submerged arc welding on the fatigue strength of parts and showed that by suitable heat and mechanical treatment the strength can be partly restored.—S. K.

On the Relation between the Actual Grain Size and Fatigue Strength of Axle Steels for Rolling Stock. R. Inoue and S. Tsujimoto. (*Sumi. Met.*, 1957, **9**, Oct., 216–220). Fatigue strength is unaffected by grain size in the range 2·5–7·0 in both plain and notched test-pieces. Regardless of grain size, specimens having a sorbitic structure have a higher fatigue strength than those having a pearlitic structure in flat (but not in notched) test-pieces.—K. E. J.

Influence of Orientation Bands on Contact Fatigue Resistance of Hardened Steels. A. S. Shein. (*Metallov. Obra. Met.*, 1957, (12), 61–66). Under the influence of repeated load applications, cracks appear on the rolling surfaces of ball bearings resulting in fatigue erosion of the metal surface. Three directions of orientation bands in relation to the applied stresses are possible: The direction of the maximum tensile stresses (*a*) corresponds to the direction of the bands in the metal (along the specimen), (*b*) is situated in one plane together with bands but the position of the latter is transversal to the stresses (transversal specimen), (*c*) is situated in the surface plane, but the bands are oriented towards the surface (transversal specimen as in case (*b*) if turned through 90°). With a transversal orientation in tempered steel the tensile strength is 20% greater and the contact fatigue life 3 to 5 times longer than when the bands are oriented as in (*c*). This study resulted in an appreciable increase in the service life of ball bearings.—L. H.

Effect of Grinding Direction and of Abrasive Tumbling on the Fatigue Limit of Hardened Steel. L. P. Tarasov, W. S. Hyler and H. R. Letner. (*A.S.T.M. preprint*, 1958, (71), pp. 11). No directional effect on fatigue limit was found for two grinding conditions. The surface was strengthened by compressive stresses and failures nucleated below the surface and raised the fatigue limit. The transverse scratches did not reduce fatigue limits as the surface was strong enough to result in subsurface nucleation. Abrasive tumbling produced compression stresses and raised fatigue limit, with subsurface failure nucleation.

Influence of Carbon on the Fatigue Limit of Case-Hardened Steel. I. D. Rybasenko. (*Metallov. Obrabotka Met.*, 1957, (10), 19–21). Specimens of two steels respectively: 0·22, 0·44%C; 0·36, 0·34%Si; 0·44, 0·50%Mn; 0·030, 0·035%S; 0·040, 0·040%P, 1·03, 0·02%Cr; 0·25, 0·23%Ni) were subjected to case-hardening in activated wood charcoal covered with barium carbonate for 6, 13 and 15 hr. at 900–920° C, hardened, tempered and fatigue tested. With up to 1·1%C in the diffusional layer, fatigue strength increases with increasing carbon content in the unsaturated core; when there is over 1·1% in the layer the fatigue limit falls, especially with higher initial carbon content in the steel. The hardened-layer structures of both steels were identical.—S. K.

Fatigue Strength Reduction Factors for Inclusions in High Strength Steels. H. N. Cummings, F. B. Stulen and W. C. Schulte. (*WADC Tech. Report* 57–589; *ASTIA Document No.* 151162, *PB* 131816, 1958, Apr., pp. 32). Tentative values are given for non-metallic non-malleable inclusions in single-nucleus fractures of rotating beam specimens. Tests on 309 specimens of SAE 4340 and 4350 steels of 140–300 ksi U.T.S. were used. Values appear to depend upon the size of the inclusion and the hardness of the steel.

Method for Estimating Fatigue in Conveyor Traction Chains. I. G. Shtokman and S. I. Lyakovitskii. (*Vestnik Mashin.* 1958, (8), 23–26). On the basis of the total number of cyclic loads applied to one link of the traction chain during its service life as determined from the equation $N = iA$, where i is the number of runs for a link during its service life and A the number of load cycles for a single run (formulae are also given for determining A and i) a method is worked out for estimating the limit of fatigue. It leads to conclusions regarding the selection of efficient parameters for the conveyors which in turn reduce the loads on the traction links.—L. H.

A Modern Method Applied to the Control and Testing of Chains: Endurance Test. M. J. Strebelle. (*Bull. Ann., Assoc. des Industriels de Belgique*, 1958, (68), 46–50). A fatigue test carried out on chain links provided evidence of defects, in particular welding defects. The test was supplemented in some cases by radiographic examination. Some applications of the method are given.

Improving the Service Life of Steel Lifting Cables. D. Van de Moortel. (*Génie Civil*, 1958, **135**, Oct. 15, 409–417). Fatigue testing machines are described and methods of failure by corrosion are illustrated. Galvanised cables are included and currently used and special greases are tested in sea water. Life can be increased by proper choice of steel, use of wire redrawn after zincing and by regular lubrication.

Tests to Determine the Influence of Building-Up With Low-Hydrogen Electrodes on the Fatigue Strength of Medium Carbon Steel Shafts. (*Lastechniek*, 1958, **24**, July, 158–159). Disappointing fatigue stength results on built-up medium carbon steel shafts, using low-hydrogen electrodes were shown to be improved if special attention was paid to minimising weld defects by a careful technique. It was found that an endurance limit of about 70% of that of the unwelded material could be obtained.

Influence of State of Surface on the Tendency of 30KhGSA Steel to Delayed Fracture. T. K. Zilova, V. E. Sadkovskii and N. I. Demina. (*Metallov. Obrabotka Met.*, 1957, (8), 2–7). In the investigation described, the nature of the hardened surface layer produced by different methods on a steel (0·33%C, 0·94% Si, 0·84%Mn, 1%Cr, 0·014%S, 0·029%P) was studied together with the effect on mechanical properties. Both thermochemical and mechanical methods of surface hardening increased the tendency to delayed fracture, but the microhardness of a mechanically produced layer was found to decrease more rapidly with depth than those produced in salt baths. The sensitivity of bolts and pins to uneven loading and delayed fracture was sometimes increased by a factor of 2·5 when a thermochemically produced layer was present, the figure for mechanically produced layers being 1·5 for notch sensitivity and delayed fracture.—S. K.

The Development and Construction of Creep Testing Machines. P. Jehlička and R. Schönhöffer. (*Výzkumné Práce z Oboru Železa a Oceli, State Publ. Ho. Tech. Lit., Prague*, 1958, 197–207). [In Czech]. The construction of creep testing machines is surveyed. Details are given of the new design of batteries of creep testing machines developed by the authors. The furnaces were designed with special care so as to give a temperature fluctuation of less than ± 2°C at 500°C over the 4-in. gauge length of the specimens.—P. F.

Quick Determination of Creep Limit. T. Nishihara, S. Taira, K. Tanaka and M. Ohnami. (*Tech. Rep. Eng. Res. Inst. Kyoto Univ.*, 1957, **7**, Aug., (38), 65–79). [In English.] Correlation with tensile testing enables prediction of short time creep limit to be made, and a further extension to intermediate term creep and extrapolation to long term creep are also explained. Temperature dependence is also discussed and a formula is advanced for estimating creep limits at different temperatures, using results at a reference temperature. Agreement with measured values is satisfactory.

A Rapid Method of Quality Control of Steel for High-Pressure Steam Pipes. N. E. Karskii. (*Metall. Obrabotka Met.*, 1957, (3), 8–15). Methods of examining high pressure steam tubing samples for mechanical properties, including creep and ageing tests are described. Methods of micro- and carbide analysis are also described. Samples are taken every two to three years of service.—R. S.

Effect of Sigma Phase on Limiting Creep Stress of Alloys. E. E. Levin and E. M. Pivnik. (*Metall. Obrabotka Met.*, 1957, (2), 27–31). The three ferrous alloys tested were a 26/16/7 NiCrMo, a 12/19/1 NiCrMo containing 1·2%Si, 5·8%Mn, 2·1%W and 0·76%V and a 9/18/1 NiCrMo containing 1·4%Mn, 1·34% W, 0·17%Ti and 0·43%Nb. Carbon contents were 0·10, 0·08 and 0·37% respectively. The amounts, dimensions and distribution of the σ-phase in solid solution varied with temperature and period of holding. Impact strength at room temperature was the property most sensitive to sigma. At high temperature, ductile properties (especially transverse compression strength) and impact strength were fairly high. The formation of σ-phase during the limiting creep tests increases the capacity of alloys for deformation and if testing is sufficiently prolonged, the comparatively fine particles of σ-phase have scarcely any effect on the limiting creep stress.—R. S.

Theory of Transitional Creep in Metal Crystals. H. Blank. (*Arch. Eisenh.*, 1958, **29**, May, 301–308). The various segments of the creep curve are treated mathematically.

Creep Ductility Criterion for Metals. V. S. Ivanova. (*Zavods. Lab.*, 1955, **21**, (2), 212–216). "Ductility reserve" or useable elongation is advanced as a representation of the amount of creep that can occur before the third stage is reached. The relation between creep and rupture strength is established on this criterion and its application to calculations of structural components in service is described. A generalized creep diagram is produced.

Analysis and Application of Certain Creep Criteria. I. A. Oding and V. S. Ivanova. (*Vestnik Mashin.*, 1955, **35**, (5), 62–66). A "ductility reserve" criterion is proposed and a generalised diagram is constructed. The nature of these is discussed and examples of 9 steels are worked out and the use of the conceptions in design and the computation of safety factors is shown.

Investigations of the Creep Mechanism of Coarse-Grained 15/15 Nickel Chromium Steel. R. Widmer. (*Promotionsalbeit, Eidgenossischen Technischen Hochschule Zurich*, 1957, pp. 91). Systematic creep experiments were carried out on a coarse-grained austenitic steel at temperatures of 650°, 720° and 800° C, under stresses between 0·5 and 8 kg/mm². The composition of the steel selected eliminated separation phenomena as far as possible. Microscopical and X-ray fine structure examination showed a cell mechanism instead of the normal slip processes at medium and low stresses, mainly crystallographic slip occurring at high stresses. (45 refs.).

Creep Tests on Mild Steels at Room Temperature. A. Josefsson and G. Lagerberg. (*Jernkonts. Ann.*, 1958, **142**, (9), 570–584). In mainly ferritic steels a continuous creep over long periods is observed only if the ratio of C to N is reduced to very low values by addition of Ti. The influence of temperature on creep is examined. With steels containing

C or C and N in solid solution, possessing a distinct upper yield point, no creep is observed until stress exceeds the lower yield point.

Structure Changes in Austenitic Steels During Creep Tests. W. Koch, A. Schrader, A. Krisch, and H. Rohde. (**Stahl Eisen*, 1958, **78**, Sept. 4, 1251–1262). Austenitic Cr–Ni steels stabilized either with Ti or Nb—one of the latter had also 2%Mo—were creep-tested between 600 and 700° C up to 50 000 h, and lattice changes and structure changes of the carbide phases and of the intermetallic precipitates studied. The particles were isolated electrolytically and examined by microanalysis, X-ray analysis, electron diffraction and electron microscopy. Monocarbide and carbides of the type $Me_{23}C_6$ were found preferentially at the grain boundaries. Platelets of an unidentified structure were also observed at the grain boundaries. The $Me_{23}C_6$ carbides transform into monocarbides, at the same places σ-phase formation was observed.

Activation Energy of the Diffusion Creep of Metallic Non-Ordered Solid Substitution Solutions. Ya. E. Geguzin. (*Doklady A.N.*, 1958, **120**, (4), 819–822). A mathematical approach indicates three independent methods of determining the activation energy of diffusion-creep: (1) based on data of temperature dependence of the co-efficient of self-diffusion, (2) based on data of the heat of melting of the components of the solid solution, (3) experimentally, on the basis of temperature relationship of velocities of diffusion creep. (These methods were applied to Au–Ni, Cu–Ni and Pb–Sn alloys).—S. I. T.

The Threshold and Alternating Values of the Activation Energy of Creep and Other Phenomena in Solid Metals. K. A. Ossipow. (*Neue Hütte*, 1958, **3**, Oct., 619–622). Deviations from ideal crystal structure due to thermal oscillations are defined as "foci of thermodynamic instability," and it is argued that the activation of the elementary creeping process at all temperatures is effected by one and the same mechanism.—L. J. L.

Effect of Prior Creep on Mechanical Properties of Aircraft Structural Metals. II. 17-7 PH Alloy (TH 1050 Condition). J. V. Gluck, H. R. Voorkees and J. W. Freeman. (*WADC Tech. Report* 57-150; *ASTIA Document No.* 151115; *PB* 131826, 1958, Apr., pp. 91). Creep conditions were maintained for 10, 50 and 100 h unstressed or at stresses giving up to 3% deformation between 600–900° F. Tensile, compression or tension impact tests were then carried out. Loss of ductility was observed and loss of compression yield strength after creep at 600° F. Other properties showed little change apart from increase in strength.

A Summary of Compressive-Creep Characteristics of Metal Columns at Elevated Temperatures. R. L. Carlson and G. K. Manning. (*WADC Tech. Report* 57-96; *ASTIA Document No.* 151114; *PB* 131825, 1958, Apr., pp. 57). Shanley's time-dependent tangent-modulus method is found to have several practical advantages over analytical solutions. Tests were made on 2 stainless steels and on non-ferrous metals.

Stress Corrosion and Creep of High-Strength Steel Wire Used in Prestressed Concrete. W. Papsdorf. (*Stahl Eisen*, 1958, **78**, Oct. 16, 1472–1475). The author reviews recent American and German publications on stress corrosion and creep of steel wire for prestressed concrete and compares the results. Graphs illustrate the various findings. It is pointed out that as-drawn wires show a higher elongation than tempered wires if stressed below 75% of the ultimate tensile strength, above 75% this is reversed.—T. G.

Changes in Structure and Properties of 0·5% Molybdenum Tube Steels in Service. I. L. Mirkin and M. I. Solonouts. (*Metall. Obrabotka Met.*, 1957, (2), 11–18). Various Russian 0·5%Mo steels lose tensile strength due to structural changes and Mo impoverishment of the solid solution. This is particularly marked in the limiting creep stress and is directly connected with the initial structure. Deterioration in tensile properties occurs mainly during the initial period of service. Above 15 000 h, there is no marked drop in tensile properties. Data show, however, that 0·6%Mo steel preserves good resistance to temperature after 100 000 h at 480° C and when stressed at 3 t/in².—R. S.

The Measurement of Hardness. Principles, Conceptions and Methods. A. Braun. (*Schweizer Archiv.*, 1958, **24**, Apr., 106–133). The author outlines the principles of hardness measurement, discusses the relationship of hardness to other properties, the effects of heterogeneity and anisotropy, followed by a description and evaluation of the methods used.—L. D. H.

Present-Day Technique in Hardness Problems. (*Tecn. Indust.*, 1958, **36**, June, 645–648). A short description is given of various types of hardness testing instruments.

Relative Hardness and Degree of Maturity as Conceptions for Evaluating Grey Cast Iron. W. Patterson. (*Giesserei*, 1958, **45**, 3 July, 385–387). An expression is derived from McKenzie's equation (*Foundry*, 1946, **74**, 88) relating hardness ($\overline{HB}$) to tensile strength. The ratio of the measured hardness (HB) to $\overline{HB}$ is termed the relative hardness of the specimen. A similar relationship of tensile strength to degree of saturation of the iron is quoted, and the ratio of the theoretical tensile strength ($\sigma_{\overline{B}}$) to the experimental value (σ_B) is termed "degree of maturity." It is suggested that these ratios may form a basis for evaluating cast iron.

A New Hardness Number and the Main Mechanical Properties of Steel. M. S. Drozd. (*Zavodska Lab.*, 1958, (8), 1002–1007). The new hardness number is determined by indentation with a ball and is related to the yield strength of the material, to the arbitrary and to the true limits of strength and also to the true breaking stress. The modulus of stress-hardening (tan θ) is used in a newly derived quantitative relation to the new hardness number $H = F(\tan\theta)$.—L. H.

A Contribution to the Practical Application of the Micro-Hardness Testing Apparatus. L. Willing. (*Radex Rundschau*, 1958, Sept., 313–316). A definition is given of the concepts of load-dependent and load-independent micro-hardness; a development of the nomogram for the determination of micro-hardness is described that permits reading-off of the Meyer index. Some comments are made on the operation of the Reichert micro-hardness testing apparatus.—L. J. L.

Some Special Features in the Change of the Properties of 10Kh25N20 Austenitic Steel. V. I. Prosvirin and L. F. Chernov. (*Metallov. Obrabotka Met.*, 1957, (10), 5–12). Results are presented and discussed of measurements of grain size and hardness with respect to duration of heating and of hardness and other properties with respect to temperature for type 10Kh25N20 austenitic steel. The Cr, Ni and iron-contents in the carbide residues were also determined, as was the toughness in relation to duration of heating and type of heat treatment. Phase and electron-microscopical analysis indicated that maximal embrittlement on prolonged heating coincided with very considerable formation of secondary phases at grain boundaries, which is considered to be also the main cause of most property changes of the steel on heating.—S. K.

Mechanical Properties of the Hardened Layer of Cast Iron Rolls. R. Kamensky. (*Hutn. Listy*, 1958, **13**, (6), 502–505). The hardness, bending strength and elastic moduli of sections cut at various depths from the hardened layer of 6 ft long cast iron rolls of 2½ and 1·8 ft diameter were measured. The hardness is highest at the surface, declining rapidly due to coarsening of the structure on going towards the interior. In the transition layer the strength of the metal increases with increasing proportions of grey iron. The elastic modulus falls on going inwards from the surface; this effect causing a stress concentration in the hardened layer due to bending of the rolls in operation. The effect is the greater the greater the elastic modulus of the hardened layer. Alloying with Mo has a beneficial effect on the strength of the hardened layer, also lowering its elastic modulus.—P. F.

Steel Types and Quality. J. Verö. (*Kohászati Lapok*, 1958, **13**, Feb.-March, 86–91). The development trends of steel alloys, the transformation characteristics of various steels, and their influence on the hardenability are discussed.—P. K.

Austenite Homogeneity and Hardenability. A. Legat. (*Hart. Techn. Mitt.*, 1958, **12**, (3), 9–18). Factors influencing the lack of homogeneity of austenitic structures, causing variable hardenability, are discussed, some free-cutting steels being used as examples.

Strain-Ageing of High-Carbon Steels. F. de Kazinczy. (*Jernkont. Ann.*, 1958, **142**, (8), 515–535). The mechanism of improvement of elastic properties during annealing at 150–300° C was examined. The Cottrell-Bilby model of carbon diffusion was found to apply allowing for the increased solubility in α-iron at the elevated temperature. Microstresses were obtained from X-ray line broadening and were seen to decrease by about 10% during ageing. Body stresses decreased more but according to a different temperature relationship. Ageing occurs only after a considerable delay in comparison with mild steel and this increases with carbon content especially at 0·6–1·0%C and over, and also if tested in other than rolling direction. Explanations are offered. (39 refs.).

Influence of the Composition of Mild Sheet Steel on Its Toughness and Tendency to Ageing. I. E. Brainin and N. V. Gubenko. (*Metallov. Obrabotka Met.*, 1957, (9), 2–4). The results are presented of a statistical analysis of control impact-tests on 1720 sheets from 285 heats of type St.3 steel covering the following composition ranges: 0·12–0·18%C, 0·35–0·55%Mn., 0·12–0·20%Si, 0·016–0·036%P, 0·026–0·040%S. Except for Si, increases in each of the elements lowered the toughness of rolled and of mechanically aged steel but had no effect on tendency to ageing. The converse held for increasing Si content, the lesser tendency to ageing being attributed to the reduced loss of Al from the metal in deoxidation.—S. K.

Investigation of the Phenomenon of Recovery of Physical Properties During Ageing of Alloys on an Iron Base. B. G. Livshits and Van Zhun'. (*Acta. Met. Sin.*, 1958, **3**, (2), 120–128). Experiments are described for investigating the recovery of physical properties of Fe–Mo; Fe–W; Fe–Ni–Nb; and the steel N36KhT, metallographically and by X-ray analysis. It was found that the changes of properties of Fe–Mo and Fe–W alloys differ from those of Fe–Ni–Nb and N36KhT steel in that in the former the changes in ageing are connected with segregation of the second phase whilst in the latter the first stage of ageing results in the formation of the *k*-state. Recovery takes place in alloys of Fe–Ni–Nb and in steel but not in the other two alloys, probably because of the decay of the *k*-state.

Investigation of Mechanical Ageing of Steel by Superposing Real Stress-Strain Diagrams. L. I. Gladshtein, P. I. Sokolovskii and A. V. Rudchenko. (*Zavods. Lab.*, 1958, (10), 1236–1239). The standard (GOST 5520–50) method of estimating the ageing process of steel by measuring impact strength does not give comparable quantitative results. The authors describe a method in which the specimen under test undergoes a 10% preliminary extension, then it is subject to natural ageing at room temperature for 15 days or to forced ageing at 250° C for 1 h. After that it undergoes several more extensions; in each case an elongation of 2% is imparted. After each test $\ln F_0/F_x$ is calculated, where F_0 and F_x are the areas of the specimen cross-sections before and after the test respectively. Real stresses are also calculated. Stress-strain diagrams so obtained of two specimens, one in the original state and the other after ageing, are superimposed. Some interesting conclusions are deduced, one being that the natures of natural and artificial ageing differ.

The Effects of Strain-Rate and Temper Rolling on the Strain-Ageing Characteristics of Rimmed Deep-Drawing Steel. D. H. Fisher, R. L. Carlson and W. T. Lankford. (*A.S.T.M.*

preprint, 1958, (87), pp. 11). Ageing for 29, 150 and 800 h were investigated after 1·2 and 2·3% temper-rolling extension at rates from 0·001 to 0·4 per sec. The results are discussed.

Measurement of Internal Stresses Due to Local Heating. O. Iždinsky. (*Zvaranie*, 1958, **7**, (11), 322–325). Fatigue strength of welded strap joints has been measured showing that the limit can be raised considerably by suitable local heating. This is explained as due to internal stresses and this effect is discussed.

An Experimental Investigation of Some Assumptions of the Structural Theory of Creep. V. S. Ivanova and L. K. Gordienko. (*Metall. Obra. Met.*, 2–5). If creep rate depends on dislocation density it should also be reflected in conductivity and microhardness. These were measured during creep testing. An arrangement securing constant contact area and absence of thermoelectric effects was used. The expected parallelisms were found.

Creep Properties of Austenitic Nickel–Chromium Steels Containing Niobium. W. H. Bailey, M. G. Gemmill, H. W. Kirkby, J. D. Murray, E. A. Jenkinson and A. I. Smith. (*Proc. J.M.E.*, 1957, **171**, (34), 911–917). Tests were made on 16 heats of steel for steam-pipe service and recommended stresses are shown.

A Study on 12% Chromium Heat-Resisting Steels. IX. Long-period Creep and Tempering Hardness. T. Fujita. (*Tetsu to Hagane*, 1958, **44**, Mar., 419–421).—K. E. J.

The Influence of Cold Hardening on the Creep Characteristics of Austenitic Steels. N. D. Sazonova. (*Metall. Obra. Met.*, 1958, (9), 46–49). Steels EI69 and EI257 quenched from 1180° C and steel EI395 quenched from 1200° C were used. Hardness was related to extension and creep at 625° C under 12 kg/mm² was measured. Work-hardening reduces elastic deformation during loading as well as the non-steady stage of creep. Steady state creep is reduced most intensively for lower preliminary deformation values and total deformation for a given time interval is considerably reduced.

High Temperature Creep-Rupture Properties of "Conselarc"-Melted Stainless Steel. S. Yamamoto. (*Tetsu to Hagane*, 1958, **44**, Mar., 422–424).—K. E. J.

Studies on Creep Strength of the Metals Deposited by Arc-Welding. II. J. Omori. (*Tetsu to Hagane*, 1958, **44**, Mar., 421–422). Results are given for deposited metal of the following types: mild steel, 2·25 Cr–1 Mo, and 5 Cr–0·5 Mo.—K. E. J.

Studies on Creep Properties of Chromium–Molybdenum Steels. II. Effect of Deoxidation Practice on Creep Characteristics. S. Terai (*Tetsu to Hagane*, 1958, 44, Mar., 417–419). Results are given for 1Cr–0·5Mo steels, with Al additions at various levels.—K. E. J.

Creep, Stress-Relaxation, and Metallurgical Properties of Steels for Steam Power Plant Operating with Steam Temperatures Above 950° F (510° C). A. I. Smith, E. A. Jenkinson, D. J. Armstrong and M. F. Day. (*Proc. I.M.E.*, 1957, **171**, (34), 918–942). Properties of 3 bolt steels were determined and it was found that while Mo–V and Cr–Mo–V types may just be adequate for plant working at 1050° F (565° C) new steels would be needed for higher temperatures. Long-term tests on Mo–V steel show high creep-resistance, but a tendency to rupture at relatively low ductility and by intercrystalline fractures. 2·25% Cr–Mo types showed creep resistance to be lower, but rupture ductility higher with transcrystalline fracture, also some Mo depletion of the matrix was shown. Austenitic 18–12–1–Cr–Ni–Nb steels showed high creep resistance, but again a tendency to rupture. A Mo–V cast steel showed advantage over cast 0·5%Mo steel. More advanced steels are needed and further tests are being made.

Specification for the Brinell Hardness Test of Grey Cast Iron. (*Fonderia Ital.*, 1958, **7**, Nov., 430–432). [In French]. A specification for the design and procedure of Brinell hardness tests of grey cast iron is given.

On the Hardenability of Self-Hardening Steels. S. Koshiba and K. Kiyonaga. (*Tetsu to Hagane*, 1958, **44**, Apr., 487–492). Seventeen steels were investigated as to the relationship between hardness and "half-temperature time," and showed considerable differences. Mass effects deduced from Grossman's diagram agreed with the results. Bainite is susceptible to the transformation and pptn. which lead to secondary hardening. (11 references).—K. E. J.

An Eccentric Hardenability Test for Shallow Hardening Steel. I. T. Suzuki. (*Tetsu to Hagane*, 1958, **44**, Mar., 395–397). The apparatus is described, three types of test-piece are compared, and results are given for some carbon steels.—K. E. J.

The Influence of Temperature up to 1000° C on the Microhardness of Iron and Iron Alloys. H. Schenck, E. Schmidtmann, H. Brandis and K. Winkler. (*Arch. Eisenh.*, 1958, **29**, Oct., 653–660). The influence of C, N and Mn on the hardness of iron up to 1000° C was investigated with particular reference to changes in hardness associated with allotropic transformations. A hardness-testing instrument is described which was developed for determining microhardness at high temperatures in a vacuum. The results are discussed.

Influence of Quenching Stresses on the Change of Lattice Constants in the Hardening of Iron–Copper Alloys. G. Wassermann and P. Wincierz. (*Arch. Eisenh.*, 1958, **29**, Dec., 785–792). As lattice constants are influenced by stresses as well as solid solution concentrations, measurements of changes in these constants with the object of studying precipitation processes must be made in stress-free conditions. A method of allowing for the influence of stresses is described, using as example an Fe–Cu alloy containing 2%Cu.

Hardness Conversions, Knoop Versus Vickers. L. Emond. (*Met. Prog.*, 1958, **74**, Sept., 96-B, 97). (Metal Progress Data Sheet). Three types of material were used: low-carbon and high-carbon steels and Monel metal. Preparation and treatment of the metals for the test are described.—T. G.

A New Rough Surface Hardness Test for Metals Based Upon Full Indentation of a Ball. G. U. Oppel. (*Forsch. Ing.-Wes.*, 1958, **24**, (5), 149–153). [In English]. A method of hardness testing is described, in which hardness is defined as the force required for the complete penetration of the ball into the specimen. Hence the character of the surface does not affect the results, and a rough surface is permissible.

Studies on Ultra-High Strength Steel. I. T. Okada. (*Tetsu to Hagane*, 1958, **44**, Mar., 430–432). Details are given of dilation after various tempering treatments, hardness and tensile properties as functions of tempering temp., and the effect of tempering temp. on impact and notch-tensile properties.—K. E. J.

Ageing of Soft Steel. S. Hasebe. (*Tetsu to Hagane*, 1958, **44**, Mar., 354–356). Results are given for the effect of cold-working and ageing of low-C rimmed steel and Al-killed steel, and for the loss in weight in 1% H_2SO_4 of various rimmed and killed steels quenched from 930° C as a function of tempering temp.

Brittle Fracture, Fatigue, Corrosion, Cracking—Today's Steel Problem for a Shipbuilder. N. G. Leide. (*Tek. Tidskr.*, 1958, **88**, Sept. 16, 841–845). These factors are discussed with relation to shipbuilding, their comparative importance assessed, and examples given and illustrated.

On the Brittle Failure of Steels. W. Félix and T. Geiger. (*Mét. Corr. Ind.*, 1958, **33**, Feb., 85–99). The authors describe research on brittle failure of steels by impact. The properties and structure of the fractures are studied by diffraction spectra, micro-hardness tests and by metallographic techniques.—R. P.

Brittle Fracture in Low-Carbon Steel Welds. B. S. Kasatkin. (*Avtom. Svarka*, 1958, **11**, (11), 71–79). Cold brittleness of low-carbon O. H. steel which has been subjected to hot working is to a large extent determined by the special nature of the secondary, ferritic structure. The deposited metal of welded joints is distinguished from the foregoing by its essentially primary crystalline, cast structure, with a clearly marked transcrystalline formation and a somewhat greater quantity of slag inclusions than in the base metal. The cold brittleness of a typical low-carbon steel (0·05%C, 0·19%Si, 0·11%Mn, 0·028%S and 0·015%P) was determined by notch impact strength tests between − 40° and + 40° C. In conjunction with microstructures, it was established that the fracture process has two stages. The pre-rupture stage covers plastic deformation and the formation of micro-cracks in the micro-locations below the notch; the latter form preferentially within the ferritic grains (secondary structure) along the boundary zone between sections of the weld with increased and decreased contents of segregation elements (primary structure). Spherical slag inclusions counteract the development of a microcrack in individual ferritic grains. Periodic alternation between sections with an increased and decreased content of segregation elements (alloying) promotes a reduction in the extension of a microcrack, since sections with a reduced content of alloying elements display resistance to the development of a microcrack. The second stage is the development of the original microcrack, which starts with the formation of a spreading crack in a thin surface layer of the metal situated at the bottom of the notch.—M. T.

Brittle Fracture. G. M. Boyd. (*Lastechniek*, 1958, **24**, Sept., 177–186). A paper dealing with the theory of brittle fracture with some applications to welded structures, particularly ships.

Effect of the Purity of Iron on Its Low-temperature Brittleness. S. Besnard and J. Talbot. (*Compt. Rend.*, 1958, **247**, Nov. 10, 1612–1613). Tensile curves were obtained on iron with different amounts of impurities down to liquid N_2 temperature. The test-pieces showed plastic deformation only if they were previously annealed above a critical temp. which depended on the purity. Electrolytic iron of high purity required about 850° C, traces of carbon lowered it to 700–750° and elimination of C by annealing in H_2 at 700° C restored the original properties.

Influence of Silicon and Phosphorus on the Temper Brittleness of Blackheart Malleable Cast Iron. P. Bastien, P. Azou and C. Winter. (*Giesserei*, 1958, **45**, June 5, 325–333). The temper brittleness of blackheart malleable cast iron was investigated by mechanical testing, dilatometric measurements, radiographic investigation and determination of strength at elevated temperatures. The influence of Si and P was established, due to their effect in transformation temperature. A hypothesis to account for temper-brittleness was formulated, having regard to the possibility of the formation of iron carbide at 475° C.

Temper Embrittlement in Nodular Cast Irons. G. N. J. Gilbert. (*Trans. AFS.*, 1957, **65**, 49–63, discussion 64–65). Fully annealed ferritic irons were used with ferro-alloy additions, cast in 270-lb melts and test-bars $1\frac{1}{4} \times 1\frac{1}{4} \times 8\frac{1}{4}$ in. or $1\frac{3}{4} \times 1\frac{3}{4} \times 8\frac{1}{4}$ in. were cut. A Ni–Mg alloy (17%Mg) and 0·5%Si as Fe–Si were added and annealing was for 16 h at 900° C and 49 h at 690° C, furnace-cooledf Impact tests were carried out and the effects o. Si and P were noted. Transition curves were plotted and the temperatures tabulated. The recovery of ductility on slow cooling was also traced. Tensile properties were also measured and the effects of Mo studied. The conclusions are stated.

Temper-Brittleness: Part I—Unalloyed Steels. B. R. Nijhawan. (*Nat. Metallurgical Lab., Council of Sci. & Ind. Res., Jamshedpur*, 1958, 187–199). Brittle temperature transition ranges are given and effects of N_2, O_2, Al, Ti, and V in suppressing the effect are discussed. Blue brittleness is also considered and dislocation theory is used to explain the phenomena. **Part II—Alloy Steels.** B. R. Nijhawan and A. B. Chatterjea. (200–214). The inhibiting effect of Mo in Ni–Cr steels is discussed and the effects of Al and other elements are considered, grain size is shown to be a factor and an etching process giving evidence of brittleness is used to explain the phenomena. Hypotheses of the effect are reviewed. (49 refs.).

Influence of Nitrogen on the Low Temperature Brittleness of Steel. Y. Imai and T. Ishiyaki. (215–224). A study of the effects of N_2 and elements having a chemical affinity for it, including Ti, Al and V, Si, P, Cu, Su and As all of which reduce brittleness due to nitride formation to some degree, and then their individual effects supervene at higher concentrations. Charpy tests were used. (37 refs.).

The Properties of High-Speed Steel Derived from Magnetic Iron Sand. S. Koshiba. (324–332). Work of the Yasugi Works of Hitachi Ltd. on production from high grade sand from the Izumo-Hoki area is reported. Effects of the alloying elements are shown.

Hot Shortness and Grain Boundary Structure in Steel. F. Erdmann-Jesnitzer and W. Bernhardt. (*Arch. Eisenh.*, 1958, **29**, June, 353–357). The susceptibility of high-sulphur steels to red shortness is discussed, and the results of an investigation on Armco iron specimens sulphurized at the surface reported. X-Ray investigations were compared with the results of sulphur prints, and a microscopical examination of the structure was made. It was concluded that the previous metallurgical history of the steel and the presence of tramp elements were decisive factors in the incidence of red-shortness.

Present Position of the Problem of Brittle Fracture of Mild Steel. W. Soete. (*Mémoires du Centre National de Recherches Métallurgiques, Section du Hainaut*, pp. 12; from *Acta Technica Belgica Met.*, 1957, **1**, (4), 115–126). Points discussed include the influence of temperature, the mechanism of brittle fracture, the use of the Robertson machine, the effects of residual stresses, and metallurgical and structural factors influencing brittle fracture.

Contribution to the Study of the Mechanism of Brittle Fracture of Mild Steel. J. B. Lean, J. Plateau and C. Crussard. (*IRSID, Communication aux Journees D' Automne*, 1958, Sept., pp. 18). In studying the occurrence of brittle fracture it was concluded in more severe cases that rupture was preceded by the formation of Lüders bands, and that fracture occurs when deformation in the bands exceeds a certain value. The mechanism of brittle fracture is discussed in the light of the results obtained.

The Temper-Brittleness of Steel. M. V. Kazakova and N. V. Koroleva. (*Metall. Obrabotka Met.*, 1957, (2), 39–45). The opposing views that temper brittleness is due to (*a*) precipitation of an inter-granular carbide phase and (*b*) segregation of certain elements in the surface layers of the grains are discussed. The authors observed the changes at the grain boundaries by means of the electron microscope. The development of temper-brittleness with slow cooling after tempering or with prolonged holding at 350 to 550° C is the result of the precipitation of a new phase at the grain boundaries which weakens the bond. This is not a carbide phase. Elements dissolved in α-iron enter the intergranular phase. The solubility of this phase varies considerably in the range 525–575° C. An evaluation of temper-brittleness according to the variation in impact strength at room temperature can be misleading.—R. S.

Study of the Mechanical Properties and the Brittle Fracture of Mild Steel. J. B. Lean, J. Plateau and C. Crussard. (*Compt. Rend.*, 1958, **247**, Nov. 3, 1458–1461). Tensile tests on 0·07%C steel at − 210° C of durations $1{\cdot}4 \times 10^{-5}$ and $3{\cdot}2 \times 10^{-1}$ sec are used to calculate the upper elastic limit from expressions given. The growth of Lüders bands was observed in all cases before brittle facture, and when plastic deformation exceeds a critical value.

Effect of Hot Working on Temper Brittleness of Alloy Structural Steels. E. N. Sokolkov and L. V. Smirnov. (*Metall. Obrabotka Met.*, 1957, (3), 31–35). The authors state that processes causing temper brittleness occur within the grains and not only at the grain boundaries and it is, therefore, possible that such processes are concentrated at the boundaries of the substructure of the austenite grains or at the boundaries of the α-phase grains. The effect of hot working on both reversible and irreversible types of temper brittleness show that the mechanism of these phenomena is the same although their causes may be different. It is thought that the most probable theory of the phenomenon of reversible temper brittleness is the theory of precipitation but the question as to the nature of the phases causing temper brittleness requires further research.—R.S.

Temper Brittleness of the Second Type: Thermal Ageing of Steel. K. M. Pogodina-Alekseeva. (*Metall. Obrabotka Met.*, 1957, (3), 36–40). It is shown experimentally that steel prone to temper brittleness at room temperature also becomes brittle after annealing and secondary heatings similar to the tempering of quenched steel. This indicates a similarity between the phenomenon of temper brittleness of the second type and the thermal ageing of steel. Temper brittleness cannot be revealed by tests at room temperature and the most sensitive criterion is impact at temperatures corresponding to the upper threshold of cold brittleness in the toughened condition.—R. S.

The Nature of Temper (Thermal) Brittleness. A. S. Zav'yalov, L. Ya. Gol'dshtein and M. I. Senchenko. (*Metall. Obrabotka Met.*, 1957, (4), 21–30). Temper brittleness is the result of the enrichment of the boundary zones of austenite with various impurities. Some impurities in the boundary zones are in the form of isolated phases while others are in a dissolved state. When the boundary zones are enriched with impurities, the rupture strength of these zones is reduced. As a result, the critical temperature for brittleness is raised, thus causing rupture in the boundary zones. Enrichment and impoverishment of both boundary zones and zones inside the grains are discussed.—R. S.

Effect of Carbon on the Position of the Critical Range of Cold Brittleness. A. P. Gulyaev and N. P. Neverova-Skobeleva. (*Metall. Obrabotka Met.*, 1957, (4), 17–21). It is shown sometimes that an increase in the C content intensifies the tendency to cold brittleness and renders the steel more prone to temper brittleness. In some alloyed steels, an increase in the C content lowers the temperature range for transition to the brittle state and reduces the tendency to temper brittleness.—R. S.

A Program for the Appraisal of the Tendency of Brittle Fracture in Steel Forgings. A. O. Schaefer. (*A.S.M.E.*, 1957, 57–A–262, pp. 6). A committee was set up following three failures of turbine and generator rotors. A progress report is presented with an account of future projects. Methods of testing and the search for improved methods of manufacture are described.

Temper-Brittleness of Wear-Resistant Cast Steel. R. Werner. (*Giesserei*, 1958, **45**, Sept. 11, 556–560). It is shown that, apart from careful casting, cast steel calls for suitable heat treatment. After heat treatment, cast alloy steel should preferably be quenched in water. Should this set up undesirable stresses, the castings should not be cooled to room temperature, but should be removed from the water on reaching a temperature between 100 and 200° C.—L. J. L.

Effect of Phosphorus on the Susceptibility to Temper Embrittlement of Cast Cr–Mo–V Steel. J. Chaberek and R. S. Zeno. (*Proc. A.S.T.M.*, 1957, **57**, 699–714). Cast 1% Cr–1%Mo – ¼%V basic electric steel was used with 0·008, 0·02, 0·05 and 0·08%P, isothermally aged at 900–950° F for 1000 h. V-notch Charpy transition temperatures were obtained for each. Temper embrittlement was produced at over 0·02%P.

Effect of Phosphorus and Manganese on the Temper-Brittleness of CrNi Steel. N. V. Tolstoguzov and A. D. Kramarov. (*Metall. Obrabotka Met.*, 1957, (2), 32–39). The steels examined contained 1 to 1·7%Cr, 2·8–3·34%Ni, 0·005 to 0·09%P and 0·005 to 0·019%S and approximately 0·3%C. Even small amounts of P had a marked effect on the tendency to temper brittleness. This was intensified with an increase in the Mn content.

Embrittling Tendencies of Austenitic Superheater Materials at Elevated Temperatures. J. Hoke, F. Eberle and R. D. Wylie. (*Proc. A.S.T.M.*, 1957, **57**, 821–832). Nine wrought and nine weld alloys were exposed to 1200–1500° F for times up to 20 000 h and notched-bar impact tests were carried out at room and elevated temperatures, also tensile and bend tests. All materials showed some decrease in impact properties but all wrought materials seem to retain adequate strength for superheater service. Weld metals were embrittled, those least affected being 16Cr–8Ni–2Mo, 10Cr–50Ni, Inconel and Type 304. Solution annealing after welding lessens these tendencies.

The Brittleness of Electrotechnic Plates. F. Khol. (*Hutnik*, 1957, **24**, July-Aug., 295–299). [In Polish]. The crystal structure of transformer steel plates was investigated by means of the electron microscope. The source of brittleness was attributed to some crystal formation between the grains during the process of slow cooling at 300–700° C. Increase of plastic properties of the steel can be achieved by first heating at 700° C for 1–3 h and then cooling down to 300° C at 40° C/h. 17 photographs showing the microstructure of the grains and their boundaries are included.

Temper-Brittleness of 30HGSA Steel with and without Addition of Vanadium. S. Orzechowski and C. Gawin. (*Prace Inst. Hutn.*, 1958, **10**, (1), 1–16). V reduced impact properties if tempered between 500–575° in the Cr–Si–Mn steel which is itself susceptible to temper-brittleness.

Estimating the Tendency to Brittleness of Steels under Service Conditions. Calculation of the Influence of the Determining Factors. E. M. Shevandin. (*Zavods. Lab.*, 1958, (8), 1017–1024). The tendency to brittleness of low-alloy structural steels was determined by the critical temperature of brittleness. The estimation of brittle failure in non-welded structural steels was carried out on specimens with notches reproducing the greatest stress concentrators under real conditions. For welded structural steels the specimens received a limiting angular notch. The margins of temperatures between the critical temperatures and the lowest service temperature of the actual structural steels were considered as a reserve of plasticity characterising the tendency to brittleness of these structural steels.—L. H.

Criterion of Brittleness in the Failure of Steel. T. A. Vladimirskii. (*Zavods. Lab.*, 1958, (8), 1024–1028). The relation between the critical brittle temperature and the temperature of the formation of the granulation in the fracture is related to various properties of the material and a much closer relation between the two temperatures can only be observed in exceptional cases under the influence of the form contributing to brittle failure (for example pointed notch, large width of specimen). The difference in these criteria can be explained by the fact that some of them result from processes occurring at different stages of deformation and of failure of the metal.—L. H.

Brittle Failure of Rotor Forgings. (*ASTM. S.T.P.* 231, 1958, pp. 43). **Inspection of Large Rotor Forgings.** G. E. Danner. (3–9). **Rotor Forgings for Power Equipment Manufacturers.** T. G. Foulkes. (10–11). A review of production procedure. **The work of the Task Group on Brittle Failure with Respect to Research.** A. O. Schaefer. (12–15). 21 steels were studied, the findings are summarised. Effect of Cr, Ni and Si are given. **Requirements for Turbine and Generator Rotor Forgings.** A. W. Rankin. (16–27). **Influence of Geometric Factors on Results of Ultrasonic Testing of Heavy Forgings.** H. Krainer and E. Krainer. (28–36). **The Axial Test Bore in Turbo-Generator Rotors.** K. Kreitz. (37–43).

Wear of Metals. F. T. Barwell. (*J. Inst. Met.*, 1958, **86**, Feb., 257–269). The economic importance of wear processes, conditions of wear, fretting corrosion, and applications to practical cases are discussed, illustrating the work of the Mechanical Engineering Research Laboratory.

Wear Resistance of Carbon and High Chromium Steels. D. Ya. Vishnyakov and A. G. Vinitskii. (*Metall. Obrabotka Met.*, 1957, (4), 2–9). Abrasive tests showed that the in-

tensity of wear increases three- to ninefold compared with wear under conditions of dry sliding friction and depends on the composition and structure of the steel. An increase in the amount of carbides in carbon and high-chromium steels increases their resistance to both abrasion and dry sliding friction. The effect of the quantity of carbides on wear resistance of high-chromium steels is much less when tested with abrasives that when tested under dry sliding friction conditions.

A Study of the Metallurgical Properties That Are Necessary for Satisfactory Bearing Performance and the Development of Improved Bearing Alloys for Service up to 1 000° F. G. K. Bhat and A. E. Nehrenberg. (*WADC Techn. Report* 57-343; *ASTIA Document No.* 142117; *PB* 131609 200; 1957, Nov., pp. 68). Failures of tool steels in engines have occurred and 29 steels are now examined for bearing properties and 9, including two experimental steels with 0·7%C, 4·2%Cr, 0·6%V and 5·3%Mo and 1·31%C, 4·07Cr, 4·13%V, 5·75%W and 7·87 Mo respectively were found suitable. The recommended temperatures (none suitable for 1 000° F) are given.

Abrasive Exect of Thermosetting Moulding Compounds. Wear Resistance of Steel and Chromium-Plated Steel in Plastic Moulds. A. Lundborg. (*Wear*, 1958, **2**, Nov., 141–145; from *Kunststoffe*, 1958, Jan.). A tracer study is described.

New Method for Making Magnetic Fields Visible. L. Suchow. (*J. Appl. Phys.*, 1958, **29**, Feb., 223–224). α-Fe_2O_3 crystals can be grown in borax and when these are suspended in liquids their orientation is visible owing to their plate-like form.

Magnetic Investigation of the Tempering of Nitrided Iron and Steel. V. G. Permyakov. (*Metallov. Obrabotka Met.*, 1957, (8), 15–16). With the aid of a differential magnetometer in a strong field the tempering process of a carbon steel type (U8A) was compared with those of nitrided technical iron (0·02%C) and with the nitrided steel. They were found to be similar. The irreversible change of magnetic properties on tempering quenched nitrided iron and steel at 300–550° C is explained by nitride or carbonitride changes similar to the third transformation in the tempering of hardened steel. The test pieces used were hollow cylinders 30 mm long, 3 mm external diameter and 0·4 mm wall thickness.—S. K.

A Magnetic Probe and its Application for the Testing of Ferromagnetic Materials. A. Stryk. (*Hutnik*, 1957, **24**, July-Aug., 328–332). [In Polish]. A survey of various types of magnetic probes and their theory is given. Application of the probe F (Förster, now Reutlinger NRF) for the measurement of coercive forces and magnetic anisotropy is discussed.

Device for the Measurement of Hysteresis Loss in Whole Sheets of Transformer Steel. J. Siewierski. (*Hutnik*, 1957, **24**, July-Aug., 333–334). [In Polish.] A brief survey of devices for the measurement of power losses is given. The underlying principles of a similar instrument built by IMZ (Institute of Iron Metallurgy) are outlined.

Influence of Light upon the Magnetization of Massive Iron. H. E. Stauss. (*Am. J. Phys.*, 1958, **26**, Nov., 551–552). No influence was observed on repeating the work of Hart (1900).

The Effect of Temperature on the Magnetic Properties of Nickel-Iron Alloys. J. J. Clark and J. F. Fritz. (*WADC Tech. Note* 57-434; *ASTIA Document No* 142267; *PB* 131799, 1957, Dec., pp. 33). Six proprietary alloys were examined.

The Effect of Heat Treatment in a Magnetic Field on Some Magnetic Properties of Soft Magnetic Materials. V. Havel. (*Hutn. Listy*, **13**, (7), 587–594) Experiments with cast 4·9 and 6·5%Si steel, hot-rolled 4·2%Si and cold-rolled 3%Si strip, as well as with a low-carbon steel of the "Arema"-type, showed that a 6 min heat treatment near the Curie point in fields of up to 2 Oersted appreciably increased the maximum permeability compared with materials heat-treated in the same way but without a magnetic field. The effect is not observed if the thermomagnetic heat-treatment is too short.—F. F.

Investigation into the Permeability of a few Iron–Nickel and Iron–Cobalt Alloys as Depending on Frequency, in the Range 10^5–10^7 Cycles. E. I. Kondorskii and L. G. Smirnova. (*Fiz. Met.*, 1958, **6**, (2), 237–246). Mo-Permalloy, Armco iron and Fe–Co alloys with 20, 36, 60 and 72%Co were investigated. Within the frequency range of 10^5 and 10^7 cycles a reduction in permeability was found. Theoretical and experimental values coincide and show that as a first approximation the data describe the observed relation of permeability to frequency. In the Fe–Ni and in the iron–36%Co alloys i.e. in alloys with the lowest anisotropy, the correspondence between frequency and permeability is much closer.

Coercive Force of Iron Oxide Micropowders at Low Temperatures. A. H. Morrish and L. A. K. Watt. (*J. Appl. Phys.*, 1958, **29**, July, 1029–1033).

Domain Changes during Longitudinal Magnetization of Iron Whiskers. G. G. Scott and R. V. Coleman. (*J. Appl. Phys.*, 1957, **28**, Dec., 1512–1513).

Domain Observations on Iron Whiskers. R. W. DeBlois and C. D. Graham jun. (*J. Appl. Phys.*, 1958, **29**, June, 931–939).

Single Crystal "Picture Frames" of Ferromagnetic Aluminium–Iron Alloys. R. C. Hall. (*J. Appl. Phys.*, 1957, **28**, Oct., 1212).

Magnetic Alignment of Single Iron Particles. J. H. L. Watson. (*J. Appl. Phys.*, 1957, **28**, July, 821–822). Electron micrograph study is described.

The Stress-dependent Damping Capacity of Ferromagnetic Metals. G. Sumner and K. M. Entwistle. (*JISI*, 1959, **192**, July, 238–245). [This issue].

An Apparatus for Magnetic Anisotropy Analysis. B. B. Timofeev. (**Elektrichestvo*, 1957, (5), 72–74). Apparatus for rapid non-destructive determination of anisotropy and main direction of orientation is described.

Magnetic Annealing Effect in Cobalt-Substituted Magnetite Single Crystals. R. F. Penoyer and L. R. Bickford jun. (*Phys Rev.*, 1957, **108**, Oct. 15, 271–277).

Magnetic Properties of Magnetite at Low Temperatures. N. S. Korolevskaya and A. S. Mil'ner. (*Fiz. Met.*, 1956, (3), (1), 186–188).

Production and Magnetic Applications of Soft Iron-Nickel Alloys in the U.S.S.R. W. J. Skotnikow. (*Hutnik*, 1957, **24**, July-Aug., 282–289). [In Polish]. Classification and general magnetic characteristic of soft Fe–Ni alloys produced in the U.S.S.R. such as 45H, 50H, 50HH, 65HH, 38HC, 42HC, 50HXC, 70HM, 70HXC, 79HMA etc. are given. The effect of technological factors on magnetic properties are discussed. Present trends of research in the U.S.S.R. are briefly summarized.

The Supplementary Heat Treatment of Shaped Pieces Cut from Electrotechnic Sheets. F. Khol. (*Hutnik*, 1957, **24**, July-Aug., 299–301). [In Polish]. The deterioration of magnetic properties (hysteresis losses) of transformer plates during cutting was investigated. It is suggested that heat treatment can be applied to the plates when they are cut up and omitted in the earlier stages.

The Relation between the Thickness and Loss Index of Transformer Sheet. B. Otta and J. Teindl. (*Sbornik-Ostrava*, 1957, **3**, (3), 21–24). [In Czech.] The eddy current losses are expressed by the formula $\text{Loss} = k.V.B^2f^2t^2/s$ where V is the electrical conductivity, B the induction, f the frequency of the current, t the thickness of the sheet and s the specific weight. Some applications of the formula in practice are considered.—P. F.

Magniko Type Alloys with Low Cobalt Content. L. M. Garmash, A. M. Morozova and M. S. Yanskaya. (*Metall. Obrabotka Met.*, 1957, (6), 8–10). Magniko alloys possessing high magnetic energy contain 24%Co. Attempts to replace Co by a cheaper element have been unsuccessful. Investigations were therefore undertaken with a view to reducing the Co content without affecting magnetic properties. It was found possible to use 21–22%Co in the mass production of permanent magnets without any substantial variation in magnetic properties.—R. S.

Studies on Cold-Workable Permanent Magnet Alloys. II. T. Mishima and K. Tachikawa. (*Nippon Kinzoku*, 1956, **20**, Apr., 198–202). [In Japanese]. The magnetic properties of Co–Fe–Cr–V, Co–Fe–Cr and Fe–Ni–Cr alloys are compared. During cold work the $\gamma \rightarrow \alpha$ transformation increases the flux density markedly, and severe cold work produces magnetic anisotropy. Ageing above 500° C causes an increase in coercive force, although hardness decreases. Slight cold work after ageing brings about a large increase in coercive force and magnetic anisotropy.—K. E. J.

The Development of Anisotropic Silicon Steel Sheets. R. Půlpan. (*Vyzkumne Práce z Oboru Železa a Oceli*, 1958, 83–90).

Electrotechnic Plates of Permalloy Type. J. Rusz and W. Babinski. (*Hutnik*, 1957, **24**, July-Aug., 302–304). [In Polish]. A survey of materials produced in the Institute of Experimental Production (Zaklad Produkcji Doswiadczalnej Imn) is given.

A Portable Lamination Detector for Steel Sheet. B. O. Smith, A. P. H. Jennings and A. G. Grimshaw. (*Brit. J. Appl. Phys.*, 1958, **9**, May, 191–193). An electrical method of detection is described and is successful where the lamination forms an electrical discontinuity. This is usually the case in cold-rolled strip but laminations in hot-rolled strip are usually not detected.

On the Change of Electrical Resistivity at an Elementary Slip. E. D. Shchukin, V. N. Rozhanskii and Yu. V. Goryunov. (*Doklady A.N.*, 1957, **115**, (6), 1101–1103). Experiments were carried out on monocrystals of Cd and Zn which were stretched under a constant load. The results of measurement of oscillogram traces are plotted on a $q/\delta l$ diagram which shows a band of values forming an S outline with the inflexion at $\delta l \approx 500$–1000Å, of particular interest are the paired jumps separated by a interval of 40–80 m.sec. The results are discussed in detail.—S. I. T.

Ultrasonics: Their Metallurgical Applications. A. Coinde. (*Bull. Cercle Etudes Mét.*, 1958, **7**, June, 319–342). This paper reviews the applications of ultrasonics to the detection of defects in metals. The physical aspects of the production, propagation and detection of sound waves having frequencies between 100 000 and 10 000 000 Hz are discussed. The methods and apparatus used are described briefly. Practical aspects, applications, interpretation of results and limitations are dealt with.—R. P.

The Measurement of Ultrasonic Attenuation by the Multiple Echo Method. J. Krautkrämer. (*Giess. Tech. Wiss.*, 1958, Apr., 1067–1075). The fundamentals of the method and the measuring technique are first described, together with the impulse generating equipment. A series of experiments are described with the detector plate in the 'near-range' and another series in the 'far-range.' Sources of error and the accuracy of the method are critically examined. Steel was found to require higher frequencies than cast iron.—R. J. W.

Instrumentation in Ultrasonic Flaw Detection. D. G. W. Claydon. (*J. Inst. Met.*, 1958, **86**, Feb., 241–251). Apparatus and its applications, and types of alarm and recording systems are reviewed.

Recent Methods of Ultrasonic Investigation of the Material of the Permanent Way on the S.N.C.F. System. M. Palmé. (*Rev. Met.*, 1957, **54**, Nov., 879–888). The technique, equipment and results obtained by ultrasonic railway line testing on the system of the French National Railways are described. Equipment based on multiple resonance and the echo method are used, and a continuous testing device carried by a track and road car is being developed.

Ultrasonic and Metallographic Investigation of Ladle Pivots. M. Kurek and T. Jaroszek. (*Hutnik*, 1958, **25**, Jan.-Feb., 39–44). [In Polish]. The advantages of ultrasonic investigations are discussed and results of eight tests on pivots are described. The faults discovered were later confirmed by metallographic investigations.

Ultrasonic Inspection of Turbine and Compressor Rotor Blades for Cracks and Other Flaws. J. G. Rasmussen. (*Nondest. Test.*, 1958, **16**, May-June, 228–236; from *Ingenioren*, 1958, **2**, Jan.). This is a very comprehensive

description of the theory and practice of routine inspection of assembled turbine and compressor blades of aircraft engines. Since the introduction of ultrasonic testing no compressor failures caused by broken blades have occurred. Probe design and construction is dealt with in detail and the interpretation of the oscillograms is discussed at length.—T. G.

Experiences in the Non-Destructive Testing of Circular Welded Seams on Thick-Walled Tubes. A. Dombrowski and M. Wandelt. (*B.W.K.*, 1958, **10**, Oct., 492–495). Large numbers of tests were carried out simultaneously on boiler and steam tubes, using ultrasonic, gamma and X-ray techniques. In many cases it was possible to examine the actual weld by taking samples, and some valuable information was obtained on the suitability of the different testing methods.

Nondestructive Testing. S. Elonka. (*Nondest. Test.*, 1958, **16**, May-June; from *Power*, **1958**, March). A review pointing out the main features and advantages of each of the following techniques under the headings: **X-Ray Is Old Standby; It Finds Flaws, Leaves a Record.** (116–117); **Isotope Radiography: A Practical Substitute for the X-Ray.** (118–119); **Fluoroscopy Is Fast, Low Cost, Gives You Story at a Glance.** (120–121); **Ultrasonic Testing Is Quick; There Are Three Methods.** (122–123); **Immersion Test Uses Liquid Couplant.** (124–125); **Resonance Method Measures Thickness, Spots Flaws.** (126–127); **Eddy-Current Testing Can Be Either Continuous or Automatic.** (128–129); **Magnetic-Particle Inspection Pinpoints Flaws near the Surface.** (130–131); **Black Light Makes Faults Easy to See.** (132–133); **Penetrant Tests Use Fluorescent Liquids, Dye and Black Light.** (134–135); **Here Are Some Jobs Nondestructive Testing Will Do for You.** (136–137).—T. G.

Modern Developments in the Fields of Radiography and Radiograph Intensification. H. Nassenstein. (*Schweissen Schneiden*, 1958, **5**, Sept., 376–379). Modern developments such as xerography and various methods of image intensification are described and critically compared. Particular attention is directed to solid-state intensification.

X-Ray Study of Lattice in Cold-Drawn Steel Wire. Ya. S. Umanskii and D. M. Zlatoustovskii. (*Metallov. Obra. Met.*, 1958, (3), 11–15). A ferritic rimming steel wire with 0·08%C was drawn in two stages with the final draw at a variety of rates (157·5–590 m/min) to 1·6 mm dia. Tensile values and X-ray patterns were obtained. Microstresses were found to increase continuously with reduction but the lattice was saturated with static distortions while stretching produced saturation with microstresses in addition. Increase of speed lowered stresses possibly due to partial relief by recovery, also distribution over cross-section becomes more uniform.

Tables of Interplanar Spacings Computed for the Characteristic Radiations of Copper, Molybdenum, Iron, Chromium and Cobalt. H. J. Garrett and R. E. Brocklehurst. (*WADC Techn. Report* 57–381; *ASTIA Document No. AD* 142344; *PB* 131764, 1958, Feb., pp. 725).

An Evaluation of Geiger Counter X-Ray Techniques for Measuring Stresses in Hardened Steels. K. E. Beu. (*Proc. A.S.T.M.*, 1957, **57**, 1282–1289). A "two-exposure" X-ray technique is described which gives stresses with an accuracy of $\pm$ 10 000 psi and a reproducibility of $\sim \pm$ 5 000 psi on steels in the range 25–62 Rockwell C. This is the only reliable method so far found.

Attenuation of X-Ray Reflection from α-Iron on Account of Extinction. V. A. Il'ina, V. K. Kritskaya and G. V. Kurdyumov. (*Fiz. Met.*, 1957, **5**, (2), 381–383).

Changes in Absolute Intensity of X-Ray Interference Patterns of Cold-Worked Iron. V. A. Il'ina, V. K. Kritskaya and G. V. Kurdyumov. (*Fiz. Met.*, 1957, **5**, (2), 379–381).

Radiographic Investigations of Stacking Defects in Cold-Worked α-Iron. C. N. J. Wagner. (*Arch. Eisenh.*, 1958, **29**, Aug., 489–493). The position and form of the intensity distribution curves of some Debye-Scherrer lines from iron filings were measured; they showed a symmetrical broadening as a result of cold-working. A relationship established between particle extension and crystallographic orientation is explained by the occurrence of stacking defects.

Recommended Practice for the Radiographic Examination of Fusion-Welded Joints on Steel Sheet less than 50 mm Thick. I. I. W. Commission V and J. Moravia. (*Rev. Soudure*, 1958, **14**, (3), 141–145). General methods for the X-ray and γ-ray examination of welds and also a special, more sensitive method are reviewed and compared. Recommendations are made but these are not meant to be standard tests.—R. P.

Testing, Planning and Preparation for Studies of Macro-Structure with Gamma Rays. E. Praceus. (*Technik.*, 1958, **13**, Aug., 544–550). The application of γ-rays in nondestructive testing in the mechanical engineering industry is surveyed. Methods of using radioactive isotopes such as ^{192}Ir, ^{137}Cs and ^{60}Co are described in some detail.—L. J. L.

Broad- and Narrow-Beam Attenuation of ^{192}Ir Gamma-Rays in Concrete, Steel and Lead. V. H. Ritz. (*Nondest. Test.*, 1958, **16**, May-June, 269–272). The set-up for the broad- and narrow-beam attenuation of γ-rays from an ^{192}Ir source is described and the results presented in a semi-log plot. For steel (density 480 lb/ft^3) the narrow-beam attenuation for 10% transmission was calculated from theory and is in excellent agreement with experiment.—T. G.

The Application of Radioactive Isotopes in the Metallurgical Industry. L. Jurkiewicz. (*Hutnik*, 1958, 7–8, **25**, 242–250). [In Polish.] An elementary introduction to nuclear physics is given; certain methods of obtaining isotopes are explained. Tabulated data concerning various applications of isotopes in metallurgy and the isotopes used are quoted. The following applications are discussed: investigation of the wear of the blast-furnace refractory lining, the motion of materials in a blast-furnace; automatic control of sheet and plate thickness in rolling mills, phosphorus and sulphur exchange between molten steel and slag and in the influence of various factors (chemical composition, temperature etc.) on the rate of diffusion. The investigation of diffusion of various inclusions on the boundary of metal grains permits a better understanding of their microstructure. The extreme sensitivity of isotopic methods for the detection of impurities is emphasized. The problems connected with the application of new methods are pointed out.—S. S.

Experiences with Radioactive Isotopes in Ferrous Metallurgy. L. A. Shvartsman. (*Zavods. Lab.*, 1958, (8), 915–921). Such work as has been accomplished in a relatively short time has already resulted in theoretical and practical conclusions adopted by the steel industry. Already these first steps have opened up great prospects for the immediate future and substantial progress is bound to follow closely the successes of nuclear physics.—L. H.

The Use of Radioactive Isotopes in Non-destructive Testing. E. Schiebold and E. Becker. (*Technik.*, 1958, **13**, Aug. 1, 551). The sources of radiation discussed are ^{60}Co, ^{137}Cs, ^{192}Ir, and ^{170}Tm. Their application to non-destructive testing in the casting and welding industries is briefly described.—L. J. L.

Thickness Measurements with Radioactive Isotopes. H. Hart. (*Technik*, 1958, **13**, Aug., 556–559). The various methods available are reviewed. Fundamental physical principles of measuring thickness are given; applicability and advantages of the different methods are explained; methods using radioactive isotopes are classified by physical characteristics, by the methods of measuring employed, and by the systems of amplification used.—L. J. L.

^{137}Caesium as a Versatile Radiographic Tool. M. Rhoten and S. A. Wenk. (*Nondest. Test.*, 1958, **16**, May-June, 261–264). The merits of ^{137}Cs as a radiographic tool are stressed. With respect to energy levels it stands between ^{192}Ir and ^{60}Co and for an X-ray laboratory equipped with 200- or 250-kVp apparatus it will extend its working range. A number of examples of this kind are quoted. (The chemical symbol for caesium is given in the paper as Ce).—T. G.

The Best Conditions for the Use of Radioactive Caesium Cs-137 for the Radiation of Welds and Castings. M. Baimler. (*Hutn. Listy*, 1958, **13**, (9), 853–862). The use of ^{137}Cs as a source of γ-rays and the sensitivity of γ-radiographic analysis are discussed. ^{137}Cs is compared with ^{192}Ir and ^{170}Tm and its advantages for certain analyses are indicated. Radiographs of welds in bronze tubes and steel plates, and of cast steel and a cast bronze pump impeller are shown. In steel over 50 mm thick, diffuse radiation must be taken into account. Using fine grain film (Ilford type C) the sensitivity was about 2·5–2·0% of the steel thickness, and when using lead foils it was sensitive for steel thicknesses from 35–85 mm. This was its best field of application, with a sensitivity peak between 50 and 60 mm. The sensitivity was dependent to a marked extent on the blackness factor, attaining its peak at $D = 1{\cdot}5$–$2{\cdot}5$. ^{137}Cs is particularly suitable for this work where its long half-life is required, for instance in testing welds in steam ductings, where it is superior to ^{192}Ir, especially on thin sections.—M. T.

Irradiation of Steels. T. J. Heal. (*U.K. Atomic Energy Authority Industrial Group*, 1957, 19R–TN/C–651, pp. 9). Tensile, Charpy impact and bend tests were carried out on steel specimens after irradiation in a reactor. The programme is set out in this report. Only a few preliminary findings are given.

Measuring of Thermal Expansion at High Temperatures. G. Szonyi. (*Kohaszati Lapok*, 1958, **13**, Feb.-Mar., 135–138). Common dilatometers with vitreous silica tubes are not suitable for measuring the expansion of pure metals, alloys, and refractories at temperatures higher than 1200° C. This is because above 1150° C the silica recrystallizes, and its coefficient of dilatation becomes uncertain. But a sillimanite tube is suitable for measurements at temperatures up to 1500° C, as it neither softens, nor changes its structure.—P. K.

The Measurement of the Specific Heat of Metals at High Temperatures. F. Krauss. (*Z. Metallkunde*, 1958, **49**, July, 386–392). A general description of methods suitable for the determination of the specific heats of metals at high temperatures followed by an account of an experimental arrangement due to the author. An appendix gives previously unpublished results of measurements of the specific heat of iron at temperatures of up to 1440° C.—L. D. H.

Thermal Properties of High Temperature Materials. I. B. Fieldhouse, J. C. Hedge, J. I. Lang and T. E. Waterman. (*WADC Techn. Report* 57-487; *ASTIA Document No. AD* 150954; *PB* 131718-225, 1958, Feb., pp. 79). Stainless steels of types 17-7 PH and 446 and 6 other materials were examined. Conductivity, specific heat and coeff. of expansion were measured from 1000–3000° F.

General Conceptions Regarding Thermal Shock. Part I. P. Lapoujade. (*Silicates Ind.*, 1958, **22**, Oct., 491–496). The difficulties encountered in problems relating to the resistance of ceramics and refractories to thermal shock are discussed. Various theories regarding ceramics are analysed with a view to applying them to refractories.—R. P.

Measuring the Viscosity of Cast Iron. A. Königer. (*Giesserei*, 1958, **45**, Sept. 11, 549–556). A description is given of the development of a viscosimeter and of the mathematical principles of the method of measurement. The device can be rapidly immersed in any melt, and viscosity is read off simultaneously. Conditions in which an increase in temperature will decrease viscosity to the same extent as the addition of phosphorus have been determined. Viscosity isotherms are derived that permit calculation of the viscosity of the phosphorus-free alloy having characteristics similar to those of pure iron.—L. J. L.

Elevated-Temperature Strength of Hardenable Hot-Working Steels. K. Bungardt and O. Mülders. (*Stahl Eisen*, 1958, **78**, Aug. 7, 1119–1126). Studies of the effect of tempering temperature and time on the tensile strength of hot-working steels up to 700° C show that ultimate tensile strength, elongation,

proof stress and reduction of area after fracture depend on test temperature (T) and time (t) according to the equation: $P = T(19 + \log t)$, where P is a parameter and T is in °K. Ultimate tensile strength and proof stress fall steadily with increasing temperature of testing and tempering. The results are presented in many graphs.—T. G.

An Investigation of Three Ferritic Steels for High-Temperature Application. A. P. Coldren and J. W. Freeman. (*WADC Tech. Rep.*, 57–40; *ASTIA Document* AD118204, 1957, Apr., pp. 107). SAE 4340 Ni–Cr–Mo and "17–22–A" S and "17–22–A" V Cr–Mo–V steels were examined from 700–1100° F tempered to 300 and to 350 Brinell. A *TTT*-diagram for the last-named steel is given. Some hot-working effects are reported. Generally bainitic structures are the best but the pearlites of Cr–Mo–V steels were often as strong or stronger. Tempered martensite was often the weakest.

The Elevated Temperature Properties of Weld-Deposited Metal and Weldments. H. R. Voorhees and J. W. Freeman. (*A.S.T.M. S.T.P.*, 226, 1958, Jan., pp. 223). Tables and diagrams are given for welded carbon steel, welded Cr–Mo steels (9, with or without V and Nb), stainless steels (12), and super strength alloys including Inconel W and X and Hastelloy C and X and others.

Mechanical Properties of Steel at High Temperatures and with Various Types of Deformation. E. N. Moshnin and D. I. Berezhkovskii. (*Metall. Obrabotka Met.*, 1957, (4), 35–41). The resistance to tension, compression and bending of structural carbon and low alloy steels at forging temperatures of 700–1200° C is similar. Torsion resistance, however, is only 0·5 to 0·58 of the normal resistance to tension, compression and bending. The tensile strength of heat-resisting steels forged at 600–1200° C is lower than the compression and bending strengths. When the steels are hot-worked there is a marked increase in the resistance to deformation up to a degree of deformation equivalent to uniform elongation. With further increase in the degree of deformation at forging temperatures of 700–1200° C, the resistance to deformation is almost constant.—R. S.

The High Temperature Stability of Permanent Magnets of the Iron–Nickel–Aluminium System. A. G. Clegg and M. McCaig. (*Brit. J. Appl. Phys.*, 1958, **9**, May, 194–199).

Behaviour of Two Ferritic and Three Austenitic Steels at the Temperature of Liquid Air. H. Schumann. (*Technik*, 1958, **13**, June 6, 417–427). Investigations were carried out on the mechanical properties of two ferritic steels—CK15Al and 12Ni19Al and three austenitic steels—X12Mn; CrTi16·11 and X10CrNiTi 18·8 at − 180° C. Of these, only the last-named showed a satisfactory impact resistance.

Properties Affecting Suitability of 9% Nickel Steel for Low-Temperature Service. T. N. Armstrong, J. H. Gross and R. E. Brien. (*A.S.M.E. preprint*, 58–Met–3, 1958, pp. 9). Impact values are given down to − 320° F and drop-weight tests on ½ in. and 1-in. plates are reported. Welding processes are described and fatigue strength and comparisons with other pressure vessel steels are made. The effects of up to 10% plastic strain and the effect of subsequent stress relief on impact properties were also found.

Low-Temperature Properties of Type ShKh15 Steel. A. N. Istomina. (*Metallov. Obrabotka Met.*, 1957, (8), 24–27). An account is given of an investigation of the mechanical properties of cylinders 5 mm in dia. and 8 mm high of a steel (1·02%C, 1·5%Cr, 0·35%Mn, 0·29%Si, 0·24%Ni, 0·003%S, 0·003%P) at 20°, − 20°, − 40°, − 60° and − 183° C. The test pieces were previously subjected to various forms of heat treatment. Relative deformations under all-round non-uniform compression were measured and found to be much lower at − 183° C than at the other temperatures. Increasing temperatures of heating in hardening led to an increase in the maximal tangential stress, while increasing tempering temperature (from 150° to 250° C) led to a decrease.—S. K.

Properties of Forged Carbon and Alloy Steels and of Cast Steels at Low and Very Low Temperatures. J. Dobrý. (*Hutník*, 1958, **8**, (6), 194–198). Sub-zero mechanical properties of a series of structural and special high-alloy steels are surveyed, and the effect of impurities and heat treatment on the ductile-brittle transition temperature is discussed. The low-temperature notch impact value is shown to be the most important criterion of suitability in the case of steels working under pressure or other stresses at very low temperatures in chemical apparatus.—P. F.

Self-Diffusion of Iron in the Metal and at the Grain Boundaries. S. Z. Bokshtein, S. T. Kishkin and L. M. Moroz. (*Metall. Obrabotka Met.*, 1957, (2), 2–10). The nature of the process of self-diffusion of iron in the α- and γ-states is revealed by autoradiography. From 800 to 1 200° C, atom migration during the self-diffusion of iron occurs preferentially at grain boundaries irrespective of the type of crystal lattice. The coefficient of self-diffusion of γ-iron is established in relation to temperature separately for diffusion within the grains and at the grain boundaries. Differences in the nature of self-diffusion within the grains and at the boundaries must be due to peculiarities of the crystal lattice structure in the boundary zone.—R. S.

On the Relation between the Processes of Diffusion and the Reorganization of Lattices during the Decomposition of Solid Solutions in Alloys. Yu. A. Bagaryatskii and Yu. D. Tyapkin. (*Doklady A.N.*, 1957, **115**, (6), 1111–1114). The study of Ni–Ti and Ni–Cr–Ti alloys has shown that the transformation, on ageing, of monophase alloy into a two-phase alloy, takes place in two stages. At first, during the hardening process, diffusion of titanium atoms takes place in the original lattice with the formation of submicroscopical regions enriched or impoverished on Ti. During the second stage small regions appear with a different structure (γ-Ni_3Ti) coordinated with the mother solid solution. On further ageing, the embryo η-phase increases in size and acquires platy forms.

Determination of the Volume Autodiffusion Constants of α-Iron with the Help of Radioactive Isotopes. C. Leymonie and P. Lacombe. (*Rev. Mét.*, 1958, **55**, June, 524–530). A layer of isotopes ^{55}Fe and ^{59}Fe was deposited on the surfaces of very thin Armco iron platelets. By measuring periodically the activity of both sides of the platelets during a diffusion anneal under vacuum, it was possible to measure the diffusion coefficient.

Mass Spectrometer Measurements of the Diffusion Coefficient of Hydrogen in Steel in the Temperature Range of 25–90° C. R. C. Frank, D. E. Swets and D. L. Fry. (*J. Appl. Phys.*, 1958, **29**, June, 892–898). Approach to equilibrium by hydrogen supply to mild steel plates and rate of outgassing were measured. The former gave values in agreement with those obtained by others at higher temperatures, while the latter gave much lower values. Curves are shown and equations for them are given. **Ratio of the Diffusion Coefficients for the Diffusion of Hydrogen and Deuterium in Steel.** R. C. Frank, R. W. Lee and R. L. Williams. (898–900). The ratio was measured over the range 26–86° C by simultaneous introduction of mixtures of D_2 and H_2 by abrasion in the mixed oxides. The ratio was constant over this range with a value of 1·37 ± 0·02.

Influence of Preliminary Hardening on the Rate of Diffusion of Carbon in Austenite. M. A. Krishtal. (*Metallov. Obrabotka Met.*, 1957, (8), 20–21). In this work 0·5 mm thick plates of white iron containing 0·4%Cr and 4%C were decarburized in a stream of hydrogen to give a layer of pure austenite. Some of the plates were left unhardened, some were quenched once from 950° C, some twice. Diffusion coefficients were measured at 930°, 950° and 1 000° C by a metallographic method. The coefficient was found to decrease with increasing number of quenchings, this effect decreasing somewhat with rising temperature. The fall in the diffusion coefficient is attributed mainly to the formation of micro-cracks and is advanced as an explanation of the comparative ease of hardening the steel after the first time.—S. K.

Reactions of Diffusion Saturation of Steels by Metals. V. A. Illarionov. (*Metall. Obrabotka Met.*, 1957, (3), 2–8). The author reviews three possible types of surface. A theory of chemical transfer which supplements some present theories is explained. For instance, the transfer of Cu, Ni, Sn and other metals which can be reduced with iron can be explained by certain reactions given in this article.—R. S.

Diffusion in Multicomponent Metallic Systems. I. Phenomenological Theory for Substitutional Solid Solution Alloys. J. S. Kirkaldy. (*Can. J. Phys.*, 1958, **36**, July, 899–906). The generalized Fick's law for diffusion is examined to determine the conditions imposed by the second law of thermodynamics and by 'microscopic reversibility.' The resulting equations are solved for the boundary conditions corresponding to the semi-infinite diffusion couple.—A. G.

Surface Tension of Liquid Cast Iron and Its Influence on Graphite Form. R. Gautschi. (*Promotionsarbeit, Eidgenössischen Technischen Hochschule, Zurich*, 1957, pp. 91). The method of capillary depression for determining surface tension was checked for scatter and reproducibility, the effect of various protective atmospheres was also investigated, A being found the most suitable. The effect of impurities was examined. Surface tension was found to be a function of O and S content, and is one factor which determines the form of the graphite. (51 refs.).

Investigation on the Wetting of Different Materials by Molten Cast Iron. W. Patterson, D. Ammann, and S. Engler. (*Giess. Tech. Wiss.* 1958, Apr., 1037–1047). The basic concept of the wetting of materials in terms of surface tension, angle of contact and thermodynamic functions is considered, with special reference to the effect on crystallization. An investigation is described in detail in which the wetting effects of various cast irons are studied with different materials such as graphite, SiO_2, TiC. The temperature dependence of the wetting angle was studied, surface tension measured and the effect of Mg treatment investigated.—R. J. W.

Effects of Gaseous and Solid Addition Elements on Surface Tension and Contact Angle (On Graphite) of Various Iron–Carbon Alloys. J. Keverian and H. F. Taylor. (*Trans. A.F.S.*, 1957, **65**, 212–221). A sessile drop method is described and Fe–C–O, Fe–C–S and Fe–C–S–Ce were examined, surface tensions measured and microstructures shown. The surface tension of a pure C-saturated Fe at 1200° C is 1730 dynes/cm ± 6% and contact angle 121° ± 3%. C does not lower the surface tension of Fe. O and S lower and Ce raises the surface tension. (27 refs.).

Localisation and Diffusion of Hydrogen in Iron. J. Plusquellec, P. Azou and P. Bastien. (*Compt. Rend.*, 1958, **246**, June 30, 3628–3631). Hydrogen is located in the planes of the (112) faces and brittleness consequent upon this is demonstrated.

Hydrogen, Crack Initiation, and Delayed Failure in Steel. H. H. Johnson, J. G. Morlet and A. R. Troiano. (*WADC Tech. Report*, 57–262; *ASTIA Document AD* 118252, 1957, May, pp. 44; *PB* 131340). With uniform H_2 distribution, delayed failure occurs by controlled crack initiation and growth. The static fatigue limit decreased with high H_2 concentrations or sharp notches. With mild notches, cracks initiated well below the surface. An incubation period for initiation was observed and fracture stress is shown to be constant over a wide range of H_2 concentrations and applied stresses. The induction period seems to be due to diffusion to the point of initiation and the process is shown to be reversible and stress-induced.

Solubility of Nitrogen in Solid Iron in the Presence of Silicon. M. G. Frohberg and H. Graf. (*Stahl Eisen*, 1958, **78**, Sept. 4, 1266–1267). A review of recent work published in the literature dealing with the effect of addition elements on the solubility of N_2 in solid iron. The formation of silicon nitride below 705° C is pointed out and its effect on the heat-treatment of silicon iron noted. From analytical data and X-ray examination it is assumed that the silicon nitride precipi-

tated from Fe–Si–N alloys does not correspond to the stoichiometric composition Si_3N_4, but that it has a lower nitrogen content.—T. G.

Activity of Manganese in Liquid Iron–Manganese–Carbon Solutions. H. Schenck and F. Neumann. (*Arch. Eisenh.*, 1958, **29**, May, 263–267). By investigating the distribution of Mn between Ag and iron–carbon alloys it was possible to determine the activity of Mn in liquid Fe–Mn–C alloys, and to establish the energy of formation of manganese carbide. The displacement by Mn of the saturation limit of iron for carbon was re-determined.

The Effect of Copper on Martensite Breakdown and Hardenability of a 0·3%C Steel. A. S. Kenneford. (*JISI*, 1959, **192**, July, 215–217). [This issue].

The Influence of Vanadium on the Properties of Silicon Manganese Steel. Ma Chou-Chan. (*Acta Metallurgica Sinica*, 1956, **1**, (4), 367–382). The chemical composition of the steel investigated was C ~ 0·25%; Si ~ 1·0%; Mn ~ 1·0% with V additions between 0·0 and 0·55%. Investigations concerned the mechanical properties as a function of the temperature of tempering. After low temperature tempering, V increases the tensile strength and impact resistance, especially after tempering between 400 and 500°. Between 400 and 600° dispersion hardening and temper brittleness occur. Dispersion hardening increases with increased V content and the tendency to temper brittleness appears only with a V content above 0·3%.—L. B.

Effect of Silicon on the Properties of Shock-Resisting Silicon–Chromium–Tungsten Tool Steel. S. Koshiba and T. Kuno. (*Nippon Kinzoku*, 1956, **20**, Mar., 129–131). For steel containing approx. 1%Si, 1·2%Cr and 2·5%W, tensile and metallurgical properties at various temp. were determined. The optimum Si content is 0·5–1·0%; this ensures good properties after heat-treatment and imparts a fine sorbite structure by quenching followed by tempering.—K. E. J.

The Influence of the Replacement of Molybdenum by Tungsten on the Properties of Steels for Heat Treatment. S. Przegalinski and M. Bialecki. (*Prace Inst. Hutn.*, 1958, (2), 109–119). Melts of five steels, three Cr–Ni–Mo, one Cr–Mn and one Cr–Mo, were prepared in which the Mo contents (0·21, 0·25, 0·34, 0·24 and 0·21%) were replaced by up to four times these quantities of W, and a study was made of the tendency to temper brittleness and the hardenability obtained by thermal surface hardening. Results showed that replacement of Mo by W (not more than 3 times the quantity of Mo) in the Cr–Mn and Cr–Mo steels (Ni content < 0·16% and < 0·09% respectively, only) did not produce a tendency to brittle fracture, whereas in the other three steels, with much larger Ni contents, this tendency was apparent. While no influence on mechanical properties was observed, the hardenability of the steels was reduced.—M. T.

Effect of Boron on Austenite Grain Size in Steel. L. L. Pyatakova and O. D. Zhizhakina. (*Metallov. Obra. Met.*, 1958, (5), 27–30). The large effect of small amounts on grain growth on heating as-cast and as-forged specimens was confirmed. Preliminary deoxidation and degassing with Fe–Ti raises the temperature of onset. The effect of hot-working was also confirmed. Irregular grain size may be due to non-uniform B distribution at grain boundaries. To obtain fine grains, deoxidation with Al and then with Ti is recommended.

Activity of Aluminium in Liquid Ag–Al, Fe–Al, Fe–Al–C, and Fe–Al–C–Si Alloys. T. Mori. (*Suiyokwai-Shi*, 1956, **13**, Dec., 139–141). Data for the activity coefficient and interaction parameter given by Chipman and Floridis (1955) are corrected and recalculated.

The Activities of Aluminium and Iron in Iron–Aluminium Melts at 1600° C. R. D. Pehlke. (*Trans. Met. AIME.*, 1958, **212**, Aug., 486–487). The activities of Fe and Al in Fe–Al melts at 1600° C were derived and plotted, and the γ° of Fe in Al estimated at 0·05.

A Thermodynamic Calculation on the Equilibrium Constant of a Deoxidation Reaction $Al_2O_3(s) = 2Al + 3O$ in Molten Steel. H. Sawamura. (*Suiyokwai-Shi*, 1956, **13**, Dec., 142–144). The equation for the equilibrium constant of the reaction given by Chipman (1951) is revised, using new thermodynamic data given by Chipman and Floridis (1955). (21 references).—K. E. J.

The Effect of High Arsenic Content on Rail Steel Ductility. F. F. Sviridenko, A. N. Popova and M. G. Fradina. (*Stal'*, 1958, (8), 739–741). Increase from 0·13–0·25%As had little effect on strength and on the drop test.

Symposium on Production, Properties and Applications of Alloy and Special Steels, 1956, Feb. (*Nat. Metallurgical Lab. Council of Sci. & Ind. Res. Jamshedpur*, 1958, pp. 488). **Alloy Steels—Technical and Economic Background.** E. H. Bucknall. (3–10). An introduction on the nature of alloy steels and their economic advantages in service. **An Exposition of Alloy Steels for the Engineer.** J. S. Vatchagandhy. (11–24). Effects of alloying elements on hardenability and other properties and classification of engineering alloy steels are discussed with a note on temper brittleness. **Tailor-Made Alloy Constructional Steels.** B. R. Nijhawan. (25–37). Hardenability effects of various alloy elements and their distribution in various phases are discussed. High strength low-alloy types are listed. **Recent Developments in Alloy and Special Steels in Great Britain.** T. Bishop. (38–51). Types are briefly considered, including clad steels, with reference both to composition and to uses. **Low-Alloy High-Strength Steels.** V. G. Paranjpe and S. Visvanathan. (52–63). A review of compositions, properties and uses, including weldability. **Alloy and Tool Steel Industry in India.** K. N. P. Rao. (64–67). **Alloy Steel Industry in India—Historical.** B. R. Nijhawan and E. H. Bucknall. (68–73). Statistics of production. **Indian Alloy Steels for Engineering Industries.** B. R. Nijhawan. (74–109). Steels are discussed in relation to Indian resources of alloying elements and to engineering requirements. Analyses of standard German and Japanese steels are given and also of AISI grades and boron steels. The effects of B and of rare earth metal additions are considered and substitute steels are suggested and compared with official specifications. **An Investigation into the Development of Prospective Indian Austenitic Stainless Steels.** P. K. Gupte and B. R. Nijhawan. (130–140). The development of substitute steels is discussed. 16–16–1, 17–6–4 and 18–10–4 Cr–Mn–Ni steels are reviewed and welding rod compositions are given. Work done on mechanical and anti-corrosive properties is reported. **Substitution of Alloying Elements by Grain Size Control of Alloy Steels.** B. R. Nijhawan and A. B. Chatterjee. (162–171). Effects of austenitic grain size, influences of alloying elements, and heat-treatment results are reviewed. Recommendations for grain size for various purposes are given and the effects of Al are outlined. **Effects of Hydrogen on Alloy Steel Production.** K. C. Barraclough. (236–255). Methods of sampling liquid steel and determining H_2 in a vacuum apparatus were devised and average values for steel prepared by various processes were found. Plots of hydrogen variations during the basic O.H. process are shown and the part it plays in solidification is assessed. The effects produced in solid steel are then described and hair-line crack formation is discussed. **Hydrogen in Steel.** A. G. Quarrell. (256–264). "Wild" heats, embrittlement, hair-line and hard zone cracking may be due to H_2. Wild heats, blowhole formation and hair-line cracking and methods of avoiding them are considered. **The Effect of Variation of Ruling Section on the Mechanical Properties of Carburizing Steels.** J. E. Russell and B. Ludgate. (282–297). Standard and "substitute" qualities were subjected to tensile and impact tests in diameters from $\frac{1}{2}$–4 in. **Boron Steels.** T. G. Digges. (314–323). Replacement by B of Ni, Cr and Mo is discussed. Production is outlined and effects on mechanical properties and hardenability are reported. **Development of Creep-Resistant Alloy Steels.** G. F. Chatterjee. (389–400). Various formulae are derived and factors affecting creep properties are discussed. The effects of various elements are noted and a chance of developing steels resistant to creep and to oxidation containing 17–20%Mn and over 9%Cr is pointed out. Austenite retention and forgeability and machinability suggest ~ 11%Cr and 0·2–0·3C. Mo, V, Ti and W may be useful but as little W as possible is suggested as it impairs oxidation resistance and increases specific gravity. (19 refs.).

Controlled-transformation Stainless Steel. K. J. Irvine, D. T. Llewellyn, and F. B. Pickering. (*JISI*, 1959, **192**, July, 218–238). [This issue].

METALLOGRAPHY

Investigations of Non-Metallic Inclusions in Steel. J. Foryst. (*Hutnik*, 1958, **25**, 7–8), 257–262). [In Polish]. Work on methods for the determination of non-metallic inclusions in steel began in the Institute of Iron Metallurgy, Gliwice, in 1954. They are based mainly on German and Soviet publications. Electrolytic methods—Klinger-Koch and Fitterer-Lukaszewicz—Duvanova methods and the apparatus used are described. Methods of treatment of the anode residue with chlorine are also given. Methods for the determination of oxides and carbides give satisfactory results; methods for the determination of sulphides, phosphides and nitrides are still to be found. Work continues in the Institute; new apparatus for electrolysis and spectrographic methods is being devised. Some practical applications of the research for steel production are quoted.—S. S.

Isolation and Investigation of Oxidic Slag Inclusions in an Acid Open-Hearth 0·7%C Steel. N. A. Hovgard. (*Jernkont. Ann.*, 1958, **142**, (10), 589–612). Electrolytically isolated inclusions were analysed spectrographically and with X-rays. They appeared to be mainly endogenous and chiefly of SiO_2 or Al_2O_3. Most are small and are not deformed in rolling. An hypothesis for the formation of the different phases is advanced. (22 refs.).

Critical Comparison of Methods for Determination of Oxide Inclusions in Steel. V. A. Mchedlishvili. (*Zavods. Lab.*, 1958, (10), 1171–1178). The best methods for separating oxide inclusions in steel are electrolytic, the Klinger-Koch and Fitterer-Lukashevich-Duranova methods. Both give equally reliable results, but the latter is simpler since only one electrolyte is used instead of two. The main difficulty lies in separating carbides from oxides in the anode residue. The most reliable method for decomposing carbides is the treatment of the residue with purified gaseous chlorine at 250° C and distillation of the products obtained by heating them twice at temperatures varying from 250° to 700° and at a pressure of 10^{-3} mm Hg.—S. S.

Structure Formation in Alloys when the Molten Metal is Rapidly Cooled. I. V. Salli. (*Lit. Proizv.*, 1958, (7), 22–25). On the strength of experimental findings concerning the general theory of metastable systems, no new metastable phases can originate during supercooling in systems not included in the composition diagram under equilibrium conditions. The main cause of metastability appearing in binary systems of the eutectic type on rapid cooling is the supersaturation of the solid solutions of the components of the system. With an increase in the rate of cooling the saturation can increase to a certain extent above what it would be in a transformation without diffusion. The growth of the crystals of the metastable solid solution proceeds with increased speed and is governed by the loss of heat of crystallization from the crystallization front.—L. H.

Influence of Superheating on the Thermal Properties of Processed Structural Steels. I. E. Kontorovich and B. M. Voshedchenko. (*Fiz. Met.*, 1957, **5**, (2), 340–348). The relationships were investigated between the temperature of preliminary and final recrystallization and the microstructure, appearance of the fracture and mechanical properties of the low and medium carbon low-alloy structural steels 12Kh2N4A, 18KhNVA and 40KhNMA.

On the Stoichiometry of Cementite and Hägg Carbide. F. Marion and R. Faivre. (*Compt.*

Rend., 1958, **247**, Oct. 13, 1118–1120). The authors' method was used to determine combined C in the presence of free C and the exact stoichiometry of cementite, Fe_3C, and the imperfection of Hägg carbide, $Fe_3C_{1\cdot05}$ were shown.

Polygonization of Cast Metals and Alloys. B. A. Movchan. (*Doklady A. N.*, 1958, **120**, (3), 521–522). The study of copper, nickel and steel has shown that the changes in grain boundaries in cast metals which do not undergo phase transformations are due to the imperfection of the original crystalline structure, namely to the changes in primitive crystallites and dendrites.—S. I. T.

Austenite Recrystallisation in Structural Steel after Hot Plastic Deformation. S. Gorczyca. (*Hutnik*, 1958, **25**, (7–8), 269–278). [In Polish]. The kinetics of the austenite recrystallization process after hot plastic deformation has not yet been explained in the published literature. The article describes tests that were carried out in the Cracow Academy of Mining and Metallurgy. The chemical composition of the steel used was 0·36%C; 0·72%Mn; 0·29%Si; 0·031%P, 0·010%S, 1·43%Cr; 1·46%Ni; 0·23%Mo. Cylindrical specimens of $d = 10$ mm $h = 10$ mm were used. Plastic deformation was obtained mostly by forging; temperatures varied from 850° to 1050° C. Micrographs of axial sections of specimens were taken. Tests showed that the speed of recrystallization depended on temperature of forging and degree of deformation. The formula for the rate of growth of austenite grains was established; it is $V = \exp. K\left(-\frac{\theta}{RT}\right)$, where the energy of activation $\theta = 61\cdot000$ cal/gr. atom, $K = 8\cdot9 \times 10^8$ mm/sec.—S. S.

Study on the Nucleation of Graphite and the Kinetics of the Graphitisation of Tempered White Cast Irons. P. Laurent and M. Ferry. (*Fonderie*, 1958, Sept., 421–433). The authors discuss the kinetic theory of graphite formation in relation to experimental results. The number of graphite particles is greater if the white iron is hardened to a martensitic structure prior to annealing but this number does not usually depend on the austenizing temperature. Complete graphitization below the transformation temperature is possible after prolonged treatment. (44 references.).

Study of the Influence of Ageing Temperature on the Precipitation of Carbon in Pure Iron, by Measurement of Electric Resistance at Very Low Temperatures. B. Migaud. (*Compt. Rend.*, 1958, **247**, Oct. 6, 1003–1005). A two-stage process is disclosed, particularly well marked at ~ 150° C. Steel with 0·075%C introduced by cementation was used, annealed at 720° C and quenched, retaining 0·02%C in solid solution. Values for heat of activation for the two stages were found.

Influence of Alloying Elements on Transformations in Steel. S. Visvanathan. (*TISCO*, 958, **5**, Oct., 143–154). A review. The change of the γ-region in binary systems and in ternary systems, effects on carbides, on eutectoid composition, on hardenability and on heat-treatment and graphitization are considered. (24 references).

Chromium Distribution between the Phases in White Iron. I. E. Lev. (*Metallov. Obra. Met.*, 1957, (12), 49–52). Direct analysis of the ferrite and cementite of white iron shows that the Cr segregates mainly in the carbide phase and its distribution between the phases is not uniform but depends on many factors such as thickness of castings, temperatures and rate of cooling. An increase in the carbide phase results in an impoverishment of the ferrite in Cr. Isothermal tempering of white iron increases the Cr content in the cementite.

Stressed State of Austenite in the Neighbourhood of a Formed Martensite Crystal. B. Ya. Lyubov and A. L. Roitburd. (*Doklady A. N.*, 1958, **120**, (5), 1011–1014). Assuming that the martensite crystal is in the form of an elliptical cylinder, and on the application of a mathematical boundary problem and theory of the function of a complex variable, it is possible to calculate the stresses, deformations and elastic energy in the surrounding austenite.—S. I. T.

Investigation of the Kinetics of the Stepwise Transformation of Supercooled Austenite. B. A. Leont'ev, L. S. Palatnik and Ya. I. Spektor. (*Fiz. Met.*, 1957, **5**, (2), 304–317). The preliminary transformation of austenite into intermediate and martensitic compounds is greatly accelerated at more elevated temperatures in steels 5582 and ShKh15, the pearlitic transformation being retarded. In the case of steel 55C2 an accelerated transformation occurs at lower temperatures. The results are discussed.

Temperature Distribution during the Martensite Plate Formation. Z. Nishiyama, A. Tsubaki, H. Suzuki and Y. Yamada. (*J. Phys. Soc. Jap.*, 1958, **13**, Oct., 1084–1090). Calculations are made from published data on the rate of propagation of the transformation front. It is concluded that the nucleation may be treated as isothermal and also the growth if propagation is steady but if it is intermittent and its unit large the conditions are nearly adiabatic.

Diffusion in Multicomponent Metallic Systems. II. Solutions for Two-Phase Systems with Applications to Transformations in Steel. J. S. Kirkaldy. (*Can. J. Phys.*, 1958, **36**, July, 907–916). The generalized diffusion equation for multicomponent systems is solved for an advancing linear cylindrical or spherical two-phase interface. Application to the iron–carbon system shows that the austenite → ferrite + carbide transformation may be inhibited through the requirement of maintaining equilibrium at the interface and the interaction of carbon atoms with the off-diagonal gradients in the diffusion matrix. **III. The Motion of Planar Phase Interfaces.** (917–925). For an initially semi-infinite diffusion couple an exact solution for the motion of the planar interface is calculated. The similar though unrelated motion of markers in the diffusion zone of some single-phase couples is discussed together with Darkins phenomenological description for binary alloys.—A. G.

Austenite Transformation and the Greniger Troiano Technique. E. G. Ramchandran and C. Dasarathy. (*JISI*, 1959, **192**, July, 268). [This issue].

The Influence of the Parameters of Induction Heating on the Size of Austenite Grains in Chromium Steels. I. N. Kidin. (*Stal' Supplement Metallogr. and Heat Treatmt.*, 1958, 50–64). Test materials were chromium steels with a content of 1·0, 2·5 and 5%Cr and a carbon content between 0·4 and 1·0% at 0·2% intervals. The average size of the austenite grains for each composition at separate temperatures between 80° and 1050° at 50° intervals are tabulated.—L. B.

Change in the Dimensions of Blocks and Distortion of the Alpha and Gamma Phase of the Second Order During Tempering of Quenched Chromium–Nickel Steel. Z. K. Kos'ko. (*Fiz. Met.*, 1958, **6**, (3), 480–482). Investigations describe the changes in the fine crystal structure during transition of carbides from one composition to the other; the appearance of a special carbide corresponding to the great distortion of the crystal lattice of the α- and γ-phase of the second order, the segregation of the carbides accompanied by a splitting up of the blocks and a rapid reduction of distortion of the crystalline lattice of the second order of the α- and γ-solid solutions. These changes were investigated in steels with a constant carbon content of 0·7% and different Cr contents (3, 8 and 18%).

The Effect in Final Annealing, of the Rate of Heating and Cooling, on the Recrystallization Structure of Cold-Rolling Transformer Steels. V. V. Druzhinin and V. A. Koroleva. (*Stal' Supplt. Metallogr. and Heat Treatmt.*, 1958, 79–87). In rapid heating, a dense recrystallization structure can occur only within a rather narrow annealing temperature range depending on the material. This shows that the texture formation is closely related to the conditions of grain growth. The change in the density of the structure can be explained thus: recrystallization grains appear and grow during heating with regular and with random orientation. In slow heating, grain growth proceeds preferentially with orientation in respect to the deformation of structure and in the subsequent selective recrystallization, these grains absorb the grains randomly oriented. With a high rate of heating beyond certain fixed temperatures, the preference for regular orientation does not exist and grains grow with random orientation.—L. B.

The Cast Structure of High-Speed Steel. E. Ineson and G. Hoyle. (*Nat. Metallurgical Lab., Council of Sci. and Ind. Res., Jamshedpur*, 1958, 333–342). The cause and effects of coarse cellular carbide formation are discussed and suggestions for further investigation and possibly elimination of the defect are made. Stirring and vibration are noted. **Some Recent Advances in the Study of Transformations in Alloy Steels.** G. Mayer. (374–388). Effects of partial austenite transformation at one subcritical temperature on the isothermal transformation characteristics of the residual austenite at a different temperature are described. Formulæ for obtaining from composition the progress of martensite formation during rapid cooling are given. The general features of continuous-cooling transformation diagrams for alloy steels are discussed and their significance indicated. Relative hardenabilities as determined by different methods are compared. **Micro-Metallurgy of Alloy Steels.** B. R. Nijhawan. (306–313). Effects of minute additions, especially of boron and the rare earths on Mn steel are discussed.

The Mechanism of the Allotropic Transformation in High Purity Iron. E. Eichen and J. W. Spretnak. (*A.S.M. preprint* 89, 1958, pp. 16). A thermionic emission microscope was used to determine that shear mechanism is responsible for the transformation that occurs just below the Ae_3 temperature. Tests were made to determine the relationship between the degree of undercooling and the stress.—E. E. W.

The Solution of Graphite in Austenite. K. P. Bunin and A. A. Baranov. (*Metalloved. Obra. Met.*, 1958, (7), 15–18). When graphite dissolves in austenite it may do so at a greater rate than iron diffuses in to take its place thus producing porosity, which is self-healing. The mechanism of this pore filling process is considered. It is probable that self-diffusion is the mechanism.

Solid State Graphitization in Iron Carbon–Silicon Alloys. A. Taub. (*Foundry*, 1958, **86**, Oct., 82–83). Graphite rosette nucleation and growth in a hypoeutectic white cast iron are described, and illustrated by photographs.

A Metallographic Investigation on Carbide Reaction in Iron and Steel. I. On the Carbide Reaction of Chromium Steel during Isothermal Heating. T. Sato, H. Kaneko and E. Osonoi. (*Tetsu to Hagane*, 1957, **21**, May, 345–349). Carbide reactions are shown to proceed as follows. In 0·55%Cr steel, θ-phase dissolves chromium. In 4·29%Cr steel, θ-phase → γ-phase. In 13·28%Cr steel, θ-phase → γ-phase → ε-phase.—K. E. J.

The Role of Dislocations in Carbide Precipitation in α-Iron. R. H. Doremus. (*Acta Met.*, 1958, **6**, Nov., 674–679). Internal friction measurements showed that rate of precipitation of C in α-Fe increased with increasing strain, and a gradual change of the shape of the time-transformation curves was observed. It was concluded that precipitation of C at 120° C in undeformed and deformed α-Fe occurs by nucleation of Fe_3C particles in dislocation and their subsequent growth by carbide diffusion. (18 refs.).

Investigation of the Phase Transformation γ → (γ + α) in Iron–Nickel Alloys. M. I. Zakharova, I. A. Ignatova and N. A. Khatanova. (*Fiz. Met.*, 1958, **6**, (3), 475–479. This transformation in alloys with 30–32%Ni proceeds very slowly at 400° C. Cold deformation at room temperature speeds up the γ + α transformation and deformation of the alloy subjected to martensite transformation, speeds up the inverse α → γ process at — 196°. At temperatures above the martensite points the initial stages of the γ → α transformation proceed also by way of the reconstruction of

the lattice type present at the martensite transformation. The layers of the α-phase which form distribute themselves parallel to the [111] plane of the γ-phase. At a tempering heat of 400° C the coherence of the lattice is destroyed which is obvious in the initial stages of the transformation. Further growth of the α-phase nuclei occurs by diffusion and the zone of deformation spread around the nuclei in the matrix.—L. B.

On the Role of the Carbide Transformation in High Speed Steel when Heated Near to the A_1 Zone. Yu. A. Geller, E. V. Mel'nichenko and E. I. Onishchik. (*Stal' Supplt. Metallogr. and Heat Treatmt.* 1958, 132–148). The complex carbides of steel such as Fe_2W_2C are found in the metastable intermediate phase. The transition towards equilibrium conditions by long heating to the point A_1 shows the transformation of the complex tungsten carbides with the formation of the stable carbides WC (and W_2C) and carbides of the cementite type. It was shown experimentally that a proportion of the chromium and vanadium atoms replace the atoms of tungsten in the Fe_2W_2C carbides but not those of iron.—L. B.

Mechanism of Intermediate Transformation of Austenite. L. I. Kogan and R. I. Entin. (*Izvest. Akad. Nauk Otdel. Tekh.* 1958, (6), 12–18). Intermediate transformation involves a redistribution of carbon within the austenite and the formation of zones of high and low carbon content. Zones of low carbon concentration prevent martensite transformation while those of high carbon content may exhibit varying stability and precipitate cementite at various rates depending on the carbon content, the alloying elements and the transformation temperature.—R. S.

CORROSION

Corrosion of Steel in Water by Varied Ratios of Dissolved Gases. J. W. Watkins and G. W. Kincheloe. (*Corrosion*, 1958, **14**, July, 341t–344t). Ratios of CO_2 to O_2 and H_2S were varied and the effects on steel were observed over 984 h. Corrosion by O_2 and CO_2 increased almost in direct proportion to O_2 concentration. Inhibition by H_2S was observed over the whole range of concentration. The rate was nearly linear up to ~ 72 h and then increased up to ~ 140 h when it again became almost linear up to 984 h.

The Behaviour of Steels in Hydrogen Sulphide Environments. L. W. Vollmer. (*Corrosion*, 1958, **14**, July, 324t–328t). Sulphide stress cracking, especially in condensate wells has been investigated by a number of workers and a review is given of findings. Heat-treatment can minimise if not eliminate the process, and this must include tempering at a temperature above 1150° F and the load must be limited. For high-strength uses, non-ferrous alloys are advocated. Non-porous Ni plate is also effective.

A Case of Chromium Steel Corrosion. J. Sedzimir and Z. Zembura. (*Hutnik*, 1958, **25**, (7–8), 303–307). [In Polish.] The case of corrosion of steam valve spindles is described. These spindles were made of chromium steel, 0·16–0·24%C, 0·6%Mn, max 0·6%Si; 12–14%Cr. The part of the spindle surface which was in contact with the stuffing box packing was attacked. Packing consisted of asbestos mixed with graphite flakes and tallow. Tests were carried out in a steam chamber and electrolytically. It was found that an electrolytic process was the cause of the corrosion, the steel becoming the anode and graphite the cathode. Generous application of tallow can attenuate and even prevent the corrosion.

Corrosion of Arc Welds. F. Barbas. (*Usine Nouv.*, 1958, **14**, Oct. 23, 95, 97). Various protective measures are mentioned.

Studies of the Corrosion Behaviour of Non-Drying Oils and Greases, Special Consideration Being Given to Modern Lubricants and Preservatives for Metals. A. Bukowiecki. (*Schweiz. Arch., Wiss. Techn.*, 1958, **24**, Aug., 263–270). This section deals with some factors influencing the attack of non-drying oils and greases on metals, and reviews the literature on this subject.

The Influence of Iron Sulphide and Calcium Carbonate Deposits on the Corrosion of Sulphate Digesters. E. Uusitalo. (*Tek. Kemian*, 1958, (18), Nov. 15, 555–558). [In English.] Corrosion in (paper pulp) digesters using the sulphate cellulose process is irregular and various types of area can be distinguished on the surfaces. Bright areas of steel and black areas of sulphide are observed and both corrode, also brown areas of scale, liable to thicken and crack, and foaming on touching with HCl are observed and these do not corrode.

Chemical and Electrochemical Aspects of Corrosion in Nuclear Reactors. G. H. Cartledge. (*Werks. Korr.*, 1958, **9**, Aug.-Sept., 493–503). The contribution of electrochemical kinetics to the interpretation of the phenomena that occur at the boundary surface between the metal and its environment is discussed; it is shown how conditions in the reactor can effect this boundary phase; and some results of research in the Oak Ridge National laboratory are reported.—L. J. L.

Tinplate Corrosion in the Canning Industry. H. Cheftel. (*Werks. Korr.*, 1958, **9**, Oct., 630–636). A brief account of internal corrosion and certain kinds of internal attack such as sulphide staining is followed by a detailed examination of corrosion proper, as characterized by the electric action between tin and iron, the tin functioning as the anode. Thickness and porosity of tin plating play a secondary part: but the properties of the steel, the presence of cathode depolarizers and storage temperature are important.—L. J. L.

Effect of Shot Peening on Cavitation Resistance of Metals in Tests with the Magnetostrictive Oscillator. L. A. Glikman and Yu. E. Zobachev. (*Metallov. Obra. Met.*, 1957, (5), 38–41). Specimens of steel with 0·4–0·53%C, of brass, type LMts Zh 55–3–1 and of austenitic steel 1Kh18N9T were examined. The carbon steel specimens were normalized, those of austenitic steels were cooled in air, and the brass specimens made from molten brass received no thermal treatment at all. The diameter of shot varied between 0·6 to 1 mm and the operation was carried out at various speeds for periods up to 3 h. As a result of these tests, conclusions are reached that the relatively small increase in the hardness of the surface layer does not compensate for the other unfavourable consequences which result in a reduction of the resistance to corrosion.

A Fully Automatic Apparatus for Testing the Corrosive Effect of Solutions. G. Wildbrett. (*Werks. Korr.*, 1958, **9**, Aug.-Sept., 521–524). The apparatus described permits metal specimens to be dipped into four different solutions and to be dried in an air stream after dipping. The apparatus is described, and results of tests are reported.—L. J. L.

Rustproofing Oils for Processing. E. R. Slaby. (*Iron Steel Eng.*, 1958, **35**, Jan., 102–103). Petroleum-based oils with special additives can be used to protect strip coils during transit using only special hoods. Various causes of corrosion and their remedies are tabulated.—K. E. J.

Corrosion Resistance of Five Stainless Alloys in Nitric Acid Containing Chloride. I. L. Tingley. (*Corrosion*, 1958, **14**, June, 273–274). Apparatus is described and tests on stainless types 316, 304L, 309SCb and Ni-O-nel, all of which were corroded by condensing acid, and 329 which showed resistance for more than 384 h are reported.

Controlling Corrosion in Coal-Chemical Plants. C. P. Larrabee and W. L. Mathay. (*Corrosion*, 1958, **14**, Apr., 183–186). Data concerning constructional materials which resist corrosion in coal-chemical plants are presented. Coal handling equipment, coke ovens, primary coolers, ammonia saturators, tar-stills and light-oil recovery equipment are discussed together with structural steelwork. Low-alloy and copper-steels are used to reduce corrosion outdoors. Stainless steels are suitable for several of the most corrosive environments. Process control can also contribute to reduction of corrosion losses.

Prevention of Localized Corrosion in Sulphuric Acid Handling Equipment. G. A. Nelson. (*Corrosion*, 1958, **14**, Mar., 145–149). Carbon steel suffered accelerated corrosion adjacent to welds when handling sulphuric acid. The attack was in the form of deep grooves at or near the welds. It is attributed, as a result of field tests described, to differences in carbide structure arising during welding or hot-rivetting. Suggested methods of avoiding the corrosion are a normalizing heat treatment, avoidance of the formation of objectionable phases, and protective coatings.—J. F. S.

New Processes for the Protection of Buried Iron. T. Marković. (*Zastita Mat.*, 1958, **6**, Jan., 5–9). The use of inhibitors is considered. The physical nature of the soil should also be adjusted. It is shown that in a soil saturated with moisture corrosion is reduced to one-tenth of that in unsaturated earth and that the use of an inhibitor may reduce it a further 80%.

Cathodic Protection of a Cooling Water System. D. S. Tulloch. (*Elect. J.*, 1958, **161**, July 18, 157–159). Steel pipes carrying sea water at Kuwait are protected. The control system is described.

The Current-Voltage Relationship of Galvanic Anode Arrays in Cathodic Protection. L. J. Waldron and M. H. Peterson. (*Corrosion*, 1958, **14**, June, 289–294). Protection of U.S. warships is considered and impressed currents from sheet steel anodes were substituted for conventional anodes. An empirical equation is given for continuous linear arrays.

New Method of Preventing Corrosion on Ships' Hulls. (*Ind. Fin.*, 1958, **10**, June, 46). The method uses a zinc reference electrode the voltage from which is used to control the applied potential between lead anodes and the hull.

The Use of Magnesium for the External Cathodic Protection of Marine Vessels. C. F. Schrieber. (*Corrosion*, 1958, **14**, Mar., 126–130). The whole underwater area of several ships in active service at sea have been adequately protected by Mg anodes, some without paint coatings, for periods of 1–2 years. No fouling occurred during the tests. Suitable anode types and distributions are discussed as are the economics, which were favourable.

ANALYSIS

Modern Analytical Methods in the Iron and Steel Industry. B. Bagshawe. (*Proc. Congress Modern Analytical Chemistry in Industry*, 1957, *St. Andrews*, [1958] 21–33). Typical compositions and schemes for analysis are given and W, Co, Cu, B, C, Nb, Ta, H_2 determinations and various common reagents are reviewed.

Contributions to the Chemistry of Iron and Steel: The Foundations of Chemical Analysis (1953 to the Beginning of 1957). H. Malissa. (*Stahl u. Eisen*, 1958, **78**, May 1, 606–613). A review of the principal methods used in chemical analysis in iron and steel works. (306 references).—T. G.

A Simplified Determination of Microquantities of Carbon in Iron, Steel and Ferro-Alloy. H. Goto, T. Watanabe and K. Suzuki. (*Sci. Rep. Res. Inst. Tohoku Univ.*, 1958, **10**, Apr., 97–102). The CO_2 from combustion is condensed in a cold trap and liberated into a vacuum, the pressure being read. Agreement with the gravimetric method is within a mean standard deviation of 0·0005% and the time taken is 24–26 min.

Determining C, P and S in Steel: Use of Hilger Fluorite Polychromator. (*Iron Steel*, 1958, **31**, Mar., 101–102). The need for automatic analysis of C, P, and S in steel has been met by the Hilger fluorite polychromator, which is a spectrograph housed in a vacuum chamber and fitted with fluorite lenses and prisms. The instrument is described and an indication of its accuracy is given.—G. F.

Study of the Determination of Nitrogen in Basic Bessemer Iron. J. Marot. (*Centre Nat. de Recherches Metallurgiques*, 1957, (10); from *Ind. Chim. Belge*, 1957, **32**, pp. 16). A thermal treatment for cast iron which improves nitrogen titration is described. Nitride separation is also reported on. Specimens were heated 3 h at 950° C. Both wet and fusion methods of analysis were studied.

[Analytical Researches]. (*Metallografiska Institutets Forskningsverksamhet*, 1956, Jan.-Jun., Jul.-Dec., 1957, Jan.-Jun., pp. 26, 27, 29). **The Determination of Oxygen in Steel by the Aluminium Diffusion Method.** C. G. Carlsson. (**10**), 2–3, (**11**), 3–4; (**12**), 2–4. The full report has been published in Swedish with an English summary in *Jernkontorets Ann.*, (1956), **140**, 551–574. **Developing a Spark Generator for Spectrographic Analysis.** C. G. Carlsson. (**10**), 19–21; (**11**), 11–12. A full description of the apparatus was provided at the Colloquium Internationale VI held at Amsterdam in May 1956 and the paper appears in the relative Proceedings. **Spectrophotometric Analysis of Boron in Steel and Cast Iron.** L. Danielsson. ((**10**), 19–21; (**11**), 11–12). This paper concerns the transition of boron forms to boric acid, the separation of impurities from boric acid solutions, colour reactions and a rapid analysis method. **Accurate Analysis of Steel Alloys particularly by Direct Reading.** C. G. Carlsson. ((11), 3–4; (**12**), 4–5). Describes a special type of spectrograph for direct reading built at the Institute.—F. R. H.

The Sampling of Molten Steel for Oxygen Determination. W. M. Milton. (*Journal of the Metallurgical Club*, 1957–8, (10), 24–27). A device for collecting a sample into a copper sampler by suction, operating in 5 sec, and suitable for analysis by vacuum fusion is described. For oxygen alone a silica tube can be used.

Evaluation of Inert Gas Fusion Method for Rapid Determination of Oxygen in Steel. J. I. Peterson, F. A. Melnick and J. E. Steers jun. (*Anal. Chem.*, 1958, **30**, June, 1086–1089). A method giving the O_2 content in 7 min. or less is described and compared with vacuum fusion methods.

Sample Taking and Determination of Hydrogen in Steel under Works Conditions. H. Zitter and H. Krainer. **Appendix: On the Present State of the Work of the International Standard Committee.** A. Stetter. (*Arch. Eisenhütt.*, 1958, **29**, July, 401–410). A vacuum extraction apparatus for the determination of H_2 in steel is described. The effect of degassing temperature on the quantity of gas extracted is discussed. Samples can be taken by two methods: Collecting of H_2 during solidification or storage of the sample and freezing-in of the dissolved H_2 by rapid cooling. The latter method was adopted and is described in detail; three modifications were used. Sample taking in a split cooper mould and rapid water-quenching is the simplest method for use under work conditions.—T. G.

Analytical Use of the Reaction of Ferrous Salts with Violuric Acid. P. Černy. (*Coll. Czechoslovak Chem. Comm.*, 1958, **23**, Jan., 105–109). [In German].

Analytical Application of Phenolcarboxylic Acids of the Triphenylmethane Series. Determination of Vanadium in Steels Using Aluminon. I. S. Mustafin and E. A. Kashkovskaya. (*Zhur. Anal. Khim.*, 1958, **13**, (2), 215–219). Conditions are investigated under which V can be determined photometrically using aluminon, without interference from elements normally present in certain steels. pH was adjusted to 3·7–3·8, and optical density was measured at about 530 mμ. Accuracy and reproducibility were satisfactory.

Studies on Rapid Analysis of Iron and Steel by Photometry. IV. Rapid Photometric Determination of Small Amounts of Chromium and Nickel in Iron and Steel. O. Kammori. (*Nippon Kinzoku*, 1956, **20**, May, 255–258). A solution method based on diphenylcarbazide, using HF to avoid interference by Fe^{+++} is suitable for Cr contents of 0·005–0·78%, and requires 12–13 min. A method using dimethylglyoxime is suitable for Ni contents of 0·001–0·7%, and requires 10–12 min. **V. Rapid Photometric Determination of Small Amounts of Vanadium in Iron and Steel.** (May, 258–262). Various methods were studied. Optimum conditions were established for estimating V contents above 0·1%; the method requires 10 min, and is not affected by Cr up to 0·7%, Ni up to 5%, W up to 0·6%, and Mo up to 0·5%. For V contents of 0·001–0·07%, a different method is given; it requires 12–13 min, and up to 0·1% W does not interfere. (10 references).

Photocolorimetric Determination of Vanadium as the Phosphovanadotungstate Complex. Application to the Analysis of Steels. C. Foucart and C. Vandael. (*Publ. Assoc. Ing. Faculté Polytechn. Mons.*, 1958, (3), 23–30). The method can be applied to V contents between 0·02 and 0·36%, up to 1·5%Mo being tolerable. If the Cr or Ni content exceeds 3% a mercury cathode separation is necessary. The method can be used conversely for the determination of W. (20 references).—R. P.

Studies on the Determination of Alumina in Steel by the Acid Solution Method. S. Wakamatsu. (*Nippon Kinzoku*, 1956, **20**, May, 262–265). A method for routine use includes separation of Al_2O_3 from the metal using a 10% aq. solution of H_2SO_4 and 15%H_2O_2, carbonaceous materials and AlN being oxidized by $KMnO_4$. (16 references).—K. E. J.

Polarographic Determination of Vanadium in Ores. Z. Sulcek. (*Coll. Czechoslovak Chem. Comm.*, 1958, **23**, June, 1052–1055). Vanadium is separated from interfering elements by fusing the sample with sodium carbonate and determined polarographically in 0·2M KCN and 0·02M EDTA.—L. D. H.

Determination of Titanium in Steel by Extracting Its Complex with 1 : 8-Dihydroxynaphthalene. A. K. Babko and O. I. Popova. (*Ukrain. Khim. Zhur.*, 1957, **23**, (3), 376–380). The reaction of Ti with the reagent and some physico-chemical characteristics of the complex were investigated, and a photometric method developed for the determination of Ti in high Cr and Ni–Cr steels and in ferrosilicon. The amount of Ti in the samples ranged from 0·03 to 0·4%, and absorption was measured at ~ 478 mμ.

Colorimetric Determination of Titanium in High-Alloy Steels. M. Novotný. (*Hutn. Listy*, 1958, **13**, (5), 430–431). A rapid method for the determination of Ti in the presence of W, Mo, Cr, Ni, Fe, Al, Mn and V is described. It is accurate to within about 2%, and particularly suitable for use with Ni–Cr–V–Mo and other high-alloy steels.—P. F.

The Development of a Method for the Determination of Small Amounts of Aluminium in Steel. J. Jindra and F. Kühn. (*Výzkumné Práce z Oboru Železa a Oceli, State Publ. Ho. Tech. Lit., Prague*, 1958, 47–61). [In Czech]. Chemical and spectrographic methods are surveyed. Details are given of a spectrographic method of high accuracy developed by the authors. This was tested with steels containing 0·02–1·65%Al. The entire analysis of one sample takes 1–1½ h. A rotating carbon electrode is used.—P. F.

Electrolytic Determination of Lead in Iron and Steel. H. Gotô and Y. Kakita. (*Nippon Kinzoku*, 1956, **20**, Apr., 210–212). Details are given for determination of lead by electrolytic deposition as PbO_2.—K. E. J.

The Determination of the Surface Lead and Cadmium Contamination of Stainless Steel Tubing (Polarographic Method). F. Elliott. (*U.K. Atomic Energy Authority Industrial Group*, 1957, 190-AM/W-94, pp. 5).

Separation of the Chlorides of Molybdenum and Tungsten in the Presence of Phosphoric Acid. F. G. Zharovskii. (*Ukrain. Khim. Zhur.*, 1957, **23**, (6), 767–770). Chlorides of W can be extracted to an appreciable extent (approx. 30%) with diethyl ether, but not from 6N HCl or from 0·4N. H_3PO_4. Mo however under such conditions is extracted to the extent of 76%.—L. H.

Photometric Determination of Beryllium in Alloys of Ferrous Metals. L. P. Adamovich and B. V. Yutsis. (*Ukrain. Khim. Zhur.*, 1957, **23**, (6), 784–786). Photometric determination was carried out at pH 12·4. Such a pH is obtained by taking 5 ml of a previously prepared solution of metal in a measuring flask, adding 5 drops of indigo carmine (0·5% soln. in 50% alcohol) and titrating with a microburette against 2N HCl to appearance of colour, thus obtaining 5 ml of a buffer soln. containing indigo carmine. (A buffer soln. of 12·4 pH can be obtained by mixing 7 ml of 0·1N NaOH and 3 ml of a solution of 12·404 H_3BO_3 in 1 l 0·1N NaOH). The change of HCl volume is appreciable. In a 10 ml pyknometer add 5 ml of the soln. for investigation, sufficient HCl, and make up to the mark with the buffer soln. Transfer 3 ml into a flask add 4 ml of buffer and 3 ml. of toron soln. (0·0426 g/100 ml or $8 \cdot 10^{-4}$ mol/litre). Pour into the tube of a photocolorimeter "FEK" of 20 mm. length and measure the transmission of the soln. against a blank consisting of 7 ml. buffer soln. and 3 ml. toron soln. with a yellow colour filter. The transmission of the working soln. is taken as 100% since the blank soln. is much darker. "Toron" is benzenearsonic-1-azo—1—(2)-naphthol-3 : 6-disulphonic acid.—L. H.

The Use of Coulometry in Metallurgical Analysis, with Special Reference to Automation of Analytical Methods. M. Šícha. (*Hutn. Listy*, 1958, **13**, (6), 533–538). The utilization of the measurement of the total electric current required in certain chemical reactions as a means of chemical analysis is discussed. Examples, such as the determination of carbon, silicon, phosphorus etc. in steel, and reactions involving other ions, e.g. Mn, Cr, Ce, V, Ti, As etc. are discussed, and the chemical and electronic equipment required is considered. (24 references).—P. F.

Cathodic Polarization in the Deposition of Silicon, Iron and Sodium from Oxide Melts. O. A. Esin and V. A. Chechulin. (*Zhur. Fiz. Khim.*, 1958, **32**, Feb., 355–360). The polarization (η) in $CaO–Al_2O_3–SiO_2–MgO$ and $CaO–Al_2O_3–MgO–B_2O_3$ melts of SiO_2, Fe_2O_3 and Na_2O additions was measured at 1400–1500° C and found to be independent of the cathode material, to decrease on stirring and to follow the concentration equation. It was shown to be due to retarded diffusion of ferrous and ferric ions in the case of Fe. High-temperature polarography appears to be possible.

Gas Chromatographic Analysis. G. R Boreham and F. A. Marhoff. (*Gas Council Res. Commun.*, GC54, 1958, pp. 24).

Estimation of Tantalum and Niobium in Ores and Ferro-alloys. I. H. Wirtz and H. Rothmann. (*Z. Erz. Met.*, 1958, **11**, Oct., 465–470). $Ta_2O_5 + Nb_2O_5$ can be separated quantitatively from Ti and Sn in ores and ferroalloys. W and P precipitate simultaneously with Ta_2O_5 and Nb_2O_5, so that they have to be estimated and deducted from the total to find the sum $Ta_2O_5 + Nb_2O_5$. Four methods are suggested which have been tested by a number of laboratories. The methods are described in detail. (16 references).—T. G.

Rapid Analysis of Low-Carbon Ferrochromium for Silicon Content by the Thermoelectric Effect Method. M. I. Gasik and B. I. Emlin. (*Zavods. Lab.* 1958, (10), 1218–1219). The importance of rapid determination of Si content in Fe–Cr for the efficiency of the melting process is emphasized. The method proposed consists in using a thermocouple, of which one electrode is the specimen of solidified Fe–Cr and the other is made of copper. All other things being equal the e.m.f. of the thermocouple is a linear function of Si content. The most convenient tempreature is 300° C. The test takes about 2–3 min, including the time needed for cooling the molten specimen. The accuracy of the method is ± 0·05%.

The Spectrographic Determination of Residual Elements in Steel. The Excitation Index Technique. The BISRA Spectrographic Analysis Sub-Committee. (*JISI*, 1959, **192**, July, 253–256). [This issue].

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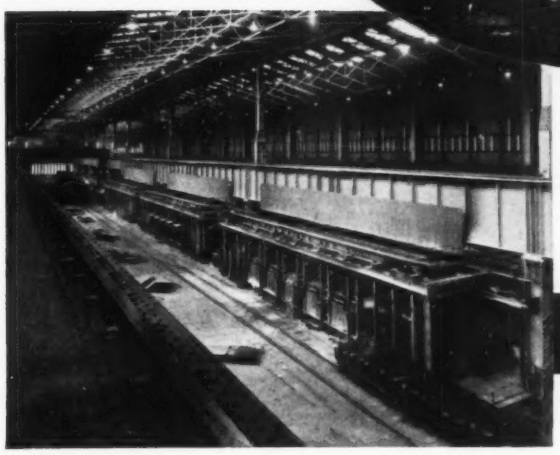

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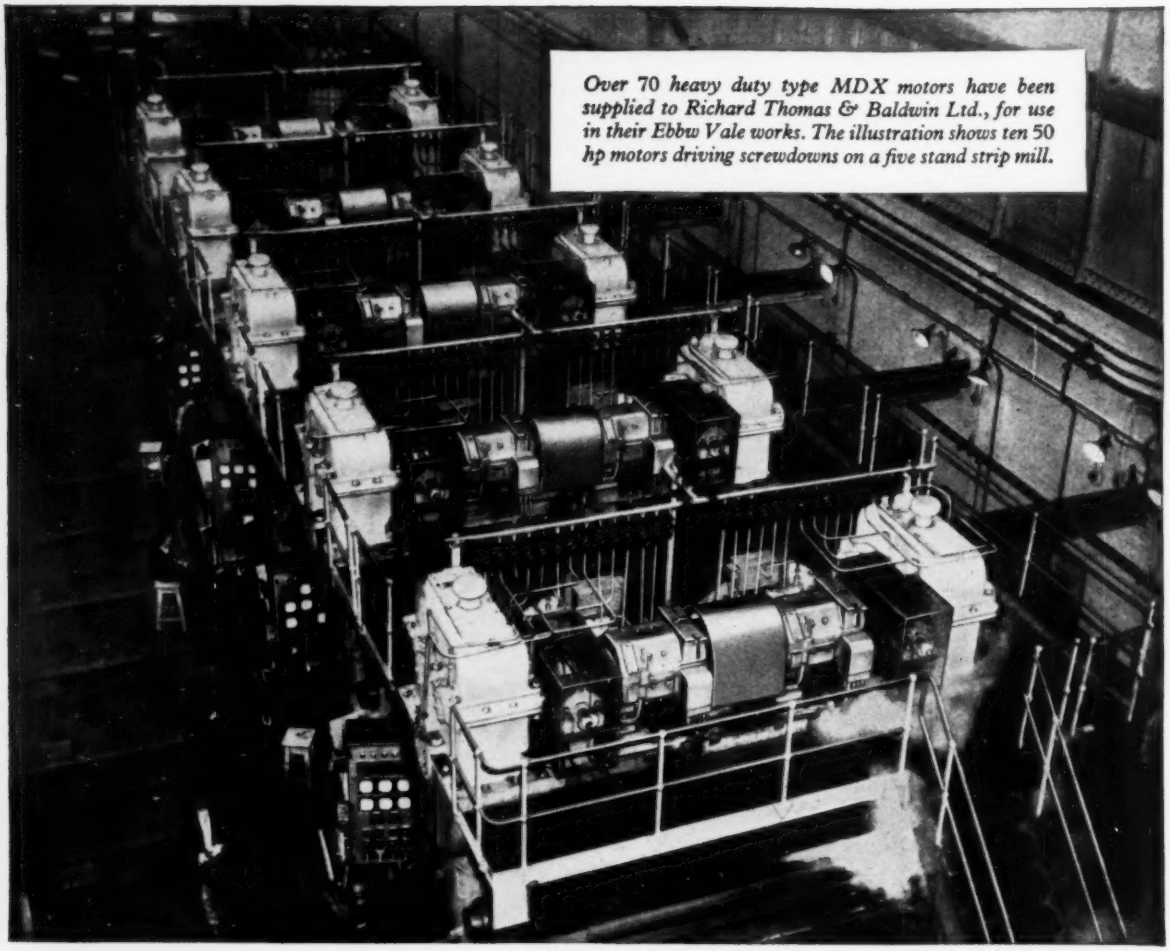

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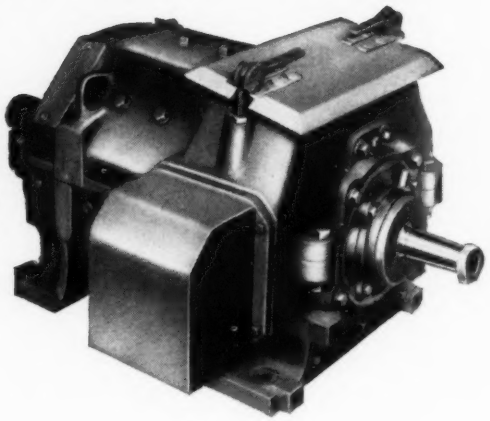

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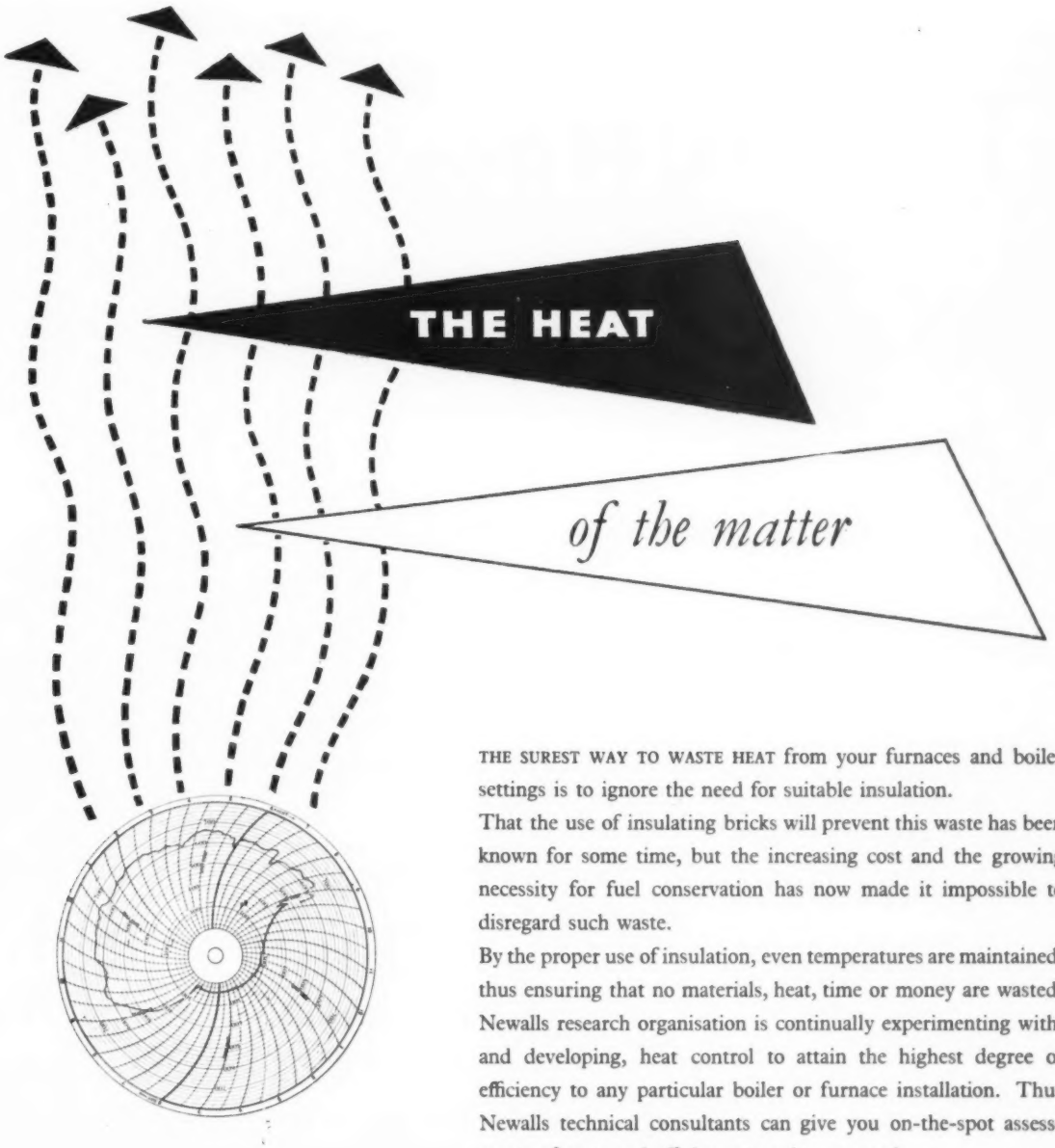

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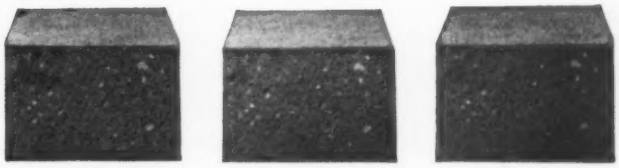

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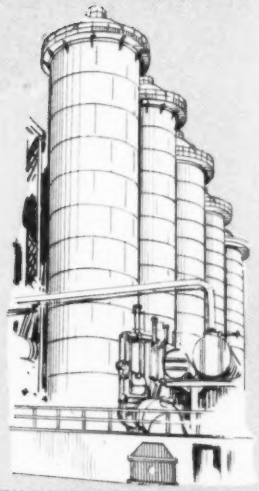

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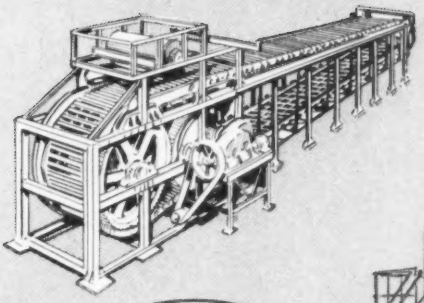

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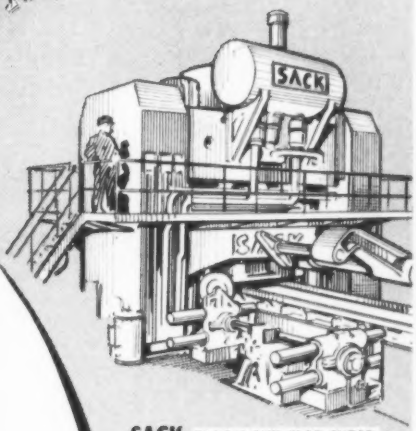

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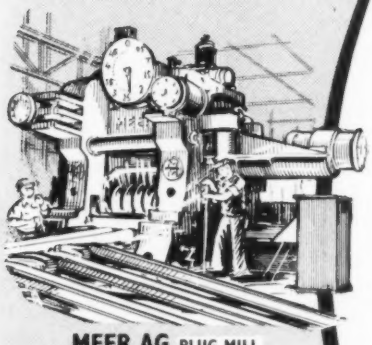

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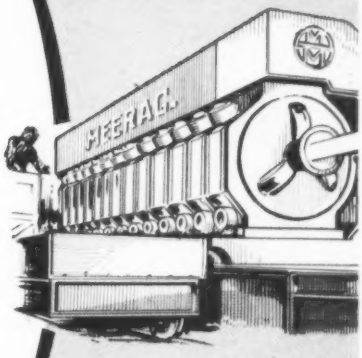

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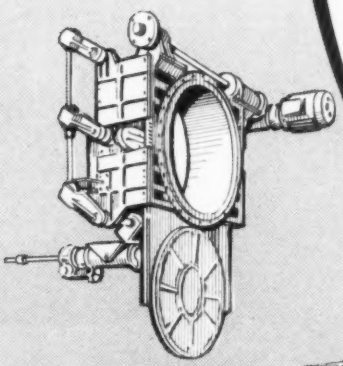

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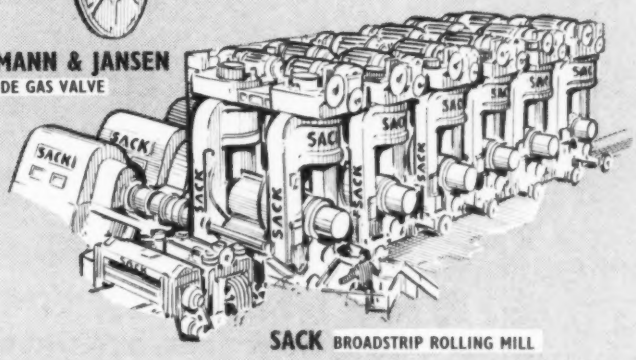

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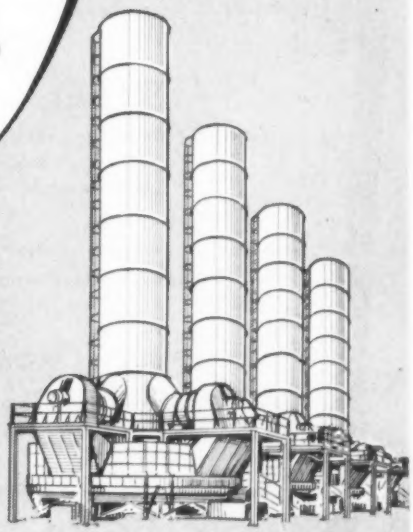

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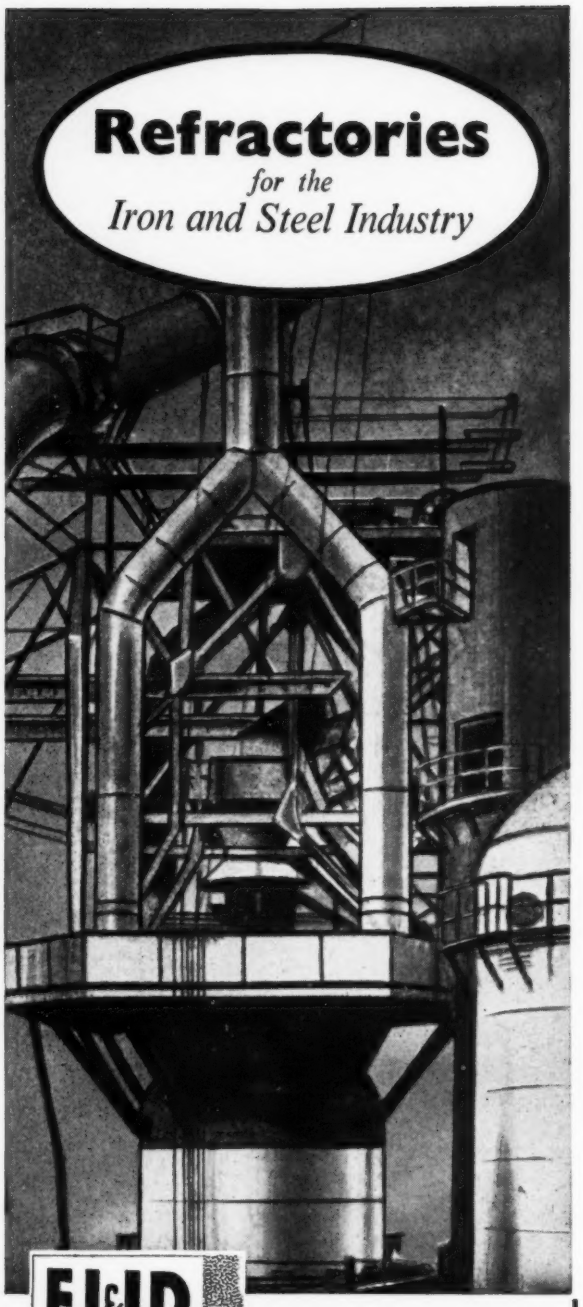

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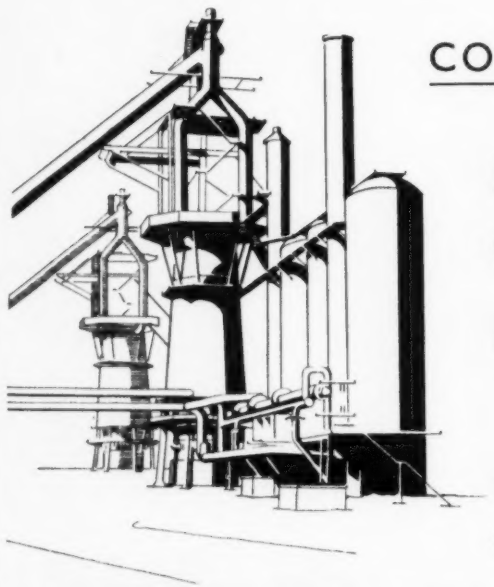

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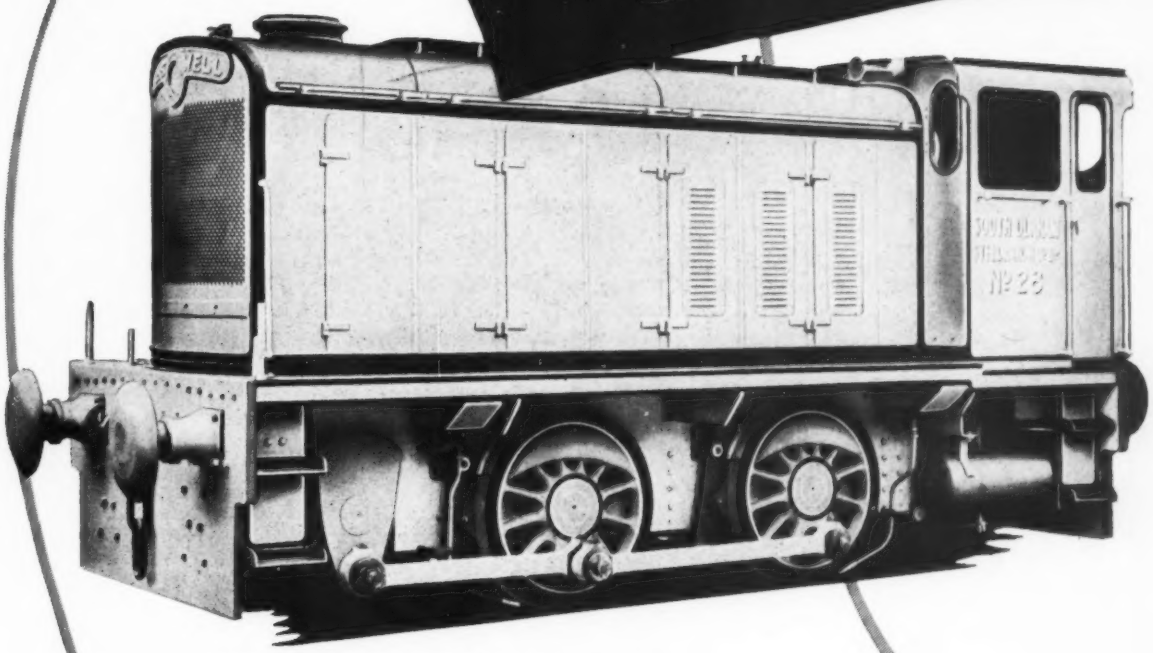

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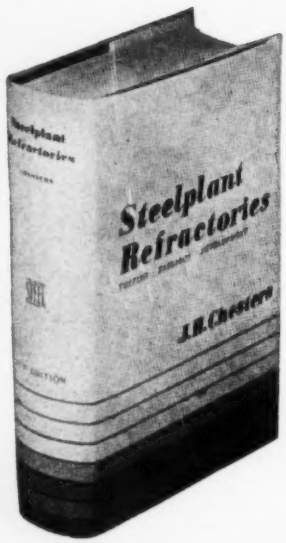

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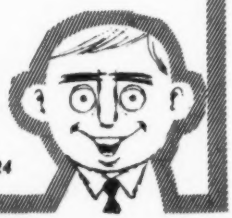

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Index to Advertisers

Adamson-Alliance Co. Ltd. 3
Adamson, Joseph, and Co., Ltd. ... ... —
Air Products (Gt. Britain) Ltd. 29
Allen West & Co., Ltd. 57
Armstrong Whitworth (Metal Industries) Ltd. ... ... —
Arrol, Sir Wm., & Co., Ltd. ... ... —
Ashmore, Benson, Pease, & Co. Inside Back Cover
Associated Electrical Industries Ltd. ... 70

Bairds & Scottish Steel, Ltd. 78
Baker Platinum Division 20
Beardmore, William, & Co., Ltd. ... ... —
Bigwood, Joshua, & Sons, Ltd. 18
Birfield Industries Ltd. 19, 59
Bonnybridge Silica & Fireclay Co., Ltd. —
Bradley & Foster Ltd. ... ... —
Brightside Foundry & Engineering Co., Ltd., The ... ... —
British Acheson Graphite Ltd. ... ... —
British Electrical Development Association —
British Furnaces, Ltd. ... ... —
British Oxygen Gases, Ltd. ... ... —
British Paints, Ltd. 68
British Rollmakers Corporation, Ltd. ... 43
British Thomson-Houston Co., Ltd., The —
British Timken, Ltd. ... ... —
Brookhirst-Igranic Ltd. 83
Brown Bayley Steels, Ltd. 52
Burn Fireclay Co. Ltd., The 84
Bushing Co., Ltd., The ... ... —

Carblox Ltd. ... ... —
Coalmoor Refractories Ltd. ... ... —
Colvilles, Ltd. 34
Consett Iron Co., Ltd. ... ... —
Constables (Matlock Quarries) Ltd. ... —
Contactor Switchgear, Ltd. ... ... —
Coppee Co. (Gt. Britain) Ltd., The ... 51
Cox & Danks, Ltd. ... ... —
Craven Bros. (Crane Division), Ltd. ... —
Crofts (Engineers) Ltd. ... ... —
Crooks, L. E. ... ... —

Davy & United Engineering Co., Ltd. ... ... Inside Front Cover
Demag 6
Dewhurst & Partner, Ltd. ... ... —
D.M.M. Machinery, Ltd. 77
Dyson, J. & J. Ltd. ... ... —

Edwards High Vacuum Ltd. ... ... —
Electric Resistance Furnace Co. Ltd. ... 85
Electroflo Meters Co. Ltd. Outside Back Cover
Engelhard Industries Ltd. 20
English Electric Co., Ltd., The ... 44, 45
English Steel Rolling Mills Corporation, Ltd. —
Evershed & Vignoles Ltd. ... ... —

Fischer Governor Co. Ltd. ... ... —
Foster, Henry, & Co., Ltd. 49
Foster Instrument Co. Ltd. ... ... —
Fraser & Chalmers ... ... —
Fuller's Earth Union Ltd., The 37

Gas Council, The ... ... —
General Electric Co., Ltd., The 41
General Refractories, Ltd. ... ... —
Gibbons Brothers, Ltd. ... ... —
Glenboig Union Fire Clay Co., Ltd. ... —
Green, E., & Son, Ltd. ... ... —
Guest Keen Iron & Steel Co., Ltd. ... 72
Guest, Keen & Nettlefolds, Ltd. 16
G.W.B. Furnaces, Ltd. ... ... —

Hadfields, Ltd. ... ... —
Harland Engineering Co., Ltd. ... ... —
Head, Wrightson, Iron & Steel Works Engineering, Ltd. 30, 31
Head, Wrightson Machine Co., Ltd. ... —
High Speed Steel Alloys Ltd. 76
Holmes, W. C. & Co., Ltd. 24
Holset Engineering Co., Ltd. The ... —
Honeywell-Controls, Ltd. 67
Howden, James & Co., Ltd. ... ... —
Hudswell Clarke & Co., Ltd. 81
Hunslet Engine Co. Ltd. 76
Huntington, Heberlein & Co., Ltd ... —

Ilford, Limited ... ... —
Imperial Chemical Industries Ltd. ... ... —
International Construction Co., Ltd., The 35

Johnson's Rolls Ltd. 17
Johnson Matthey & Co., Ltd ... ... —

Kelvin Hughes Ltd. 40
Kestner Evaporator & Engineering Co., Ltd. —
King, Geo. W., Ltd. 71
Kingscliffe Insulating Products, Ltd. ... 38

Lamberton & Co., Ltd. ... ... —
Lanarkshire Steel Co., Ltd., The ... ... —
Lancashire Steel Manufacturing Co. Ltd. ... 79
Lodge-Cottrell, Ltd. 9
Loewy Engineering Co. Ltd., The 12
Lysaght's John, Scunthorpe Works Ltd. ... 4

Marshall, Thomas, & Co. (Loxley), Ltd. ... —
Metallurgical Engineers Ltd. 53
Metropolitan Vickers Electrical Co., Ltd. ... —
Mitchell Engineering Ltd. ... ... —
Mobil Oil Co. Ltd. 61
Mond Nickel Co., Ltd. ... ... —
Moorwood, John M., Ltd. 46, 47
Morgan Crucible Co., Ltd., The ... ... —
Morgan Refractories Ltd. ... ... —
Morgardshammars Mekaniska Verkstads AB —
Moxey, Ltd. 36

National Industrial Fuel Efficiency Service ... —
Newalls Insulation Co. Ltd. 73
Newall, Ernest, & Co. Ltd. 33
Newton, Chambers & Co., Ltd. ... ... —

Oughtibridge Silica Firebrick Co. Ltd. ... —

Park Gate Iron & Steel Co., Ltd. 7
Parsons, C. A. & Co., Ltd. ... ... —
Pearson, E. J. & J., Ltd. 78
Philips Electrical Ltd. ... ... —
Phosphor Bronze Co., Ltd., The 19
Pickford Holland & Co., Ltd. ... ... —
Power-Gas Corporation Ltd., The Inside Back Cover
Priest Furnaces, Ltd. 55

Railko Ltd. 59
Rawlplug Co. Ltd., The ... ... —
Research & Control Instruments Ltd. ... —
Richardsons Westgarth & Co. Ltd. ... 10
Roberts, J. W. Ltd. 62
Robertson, W. H. A., & Co., Ltd ... ... —
Rust-Oleum Corporation 39

Salem Engineering Co., Ltd. 8
Sayvoil Co., The 15
Schloemann 28
Shaw, John, & Sons (Salford) Ltd. ... ... —
Shell-Mex & B.P., Ltd. ... ... —
Sheppard and Sons Ltd. 80
Simon-Carves, Ltd. 48
Skefko Ball Bearings Co. Ltd., The ... 1
Smith, John (Keighley), Ltd. 42
Smith & McLean, Ltd. 54
Smith, Thomas & Sons (Rodley) Ltd. ... —
Spencer-Bonecourt-Clarkson Ltd. ... 22
Stanton Ironworks Co., Ltd., The 60
Stedall & Co. Ltd. 39
Steel Co. of Wales, The 58
Stein & Atkinson, Ltd. 5
Stein, John G., & Co., Ltd. 26
Stephens Silica Brick Co., Ltd. 74
Stewarts & Lloyds, Ltd. ... ... —
Stillite Products Ltd. ... ... —
Strachan & Henshaw Ltd. 27
Summerson, Thos., & Sons, Ltd. 23
Superlite Building Materials Ltd. 84
Swann, Ratcliffe & Co. (Brassington) Ltd 2
Swinney Bros. Ltd. 82

Tattersall & Son Ltd. 80
Thermal Syndicate Ltd. 69
Thomas (Richard) & Baldwins, Ltd. ... 25
Thompson L'Hospied & Co. Ltd. ... ... —
Thornton, B., Ltd. 21

Union Carbide Ltd. ... ... —
United Coke & Chemicals Co. Ltd. ... —
United Steel Companies Ltd. 82

Vaughan, Edgar, & Co., Ltd. 13
Visco Engineering Co., Ltd. 14

Ward, Thos., W., Ltd. 75
Warner & Co., Ltd. ... ... —
Wellman Smith Owen Engineering Corporation Ltd., The 66
West Hunwick Silica & Firebrick Co., Ltd. 32
Whitehead Iron & Steel Co., Ltd. 50
Wild Barfield Electric Furnaces Ltd. ... 56
Willan, G. L. Ltd. ... ... —
Wincott, G. P., Ltd. ... ... —
Woodall-Duckham Construction Co., Ltd. 64
Wragg, Thos., & Sons Ltd. 11

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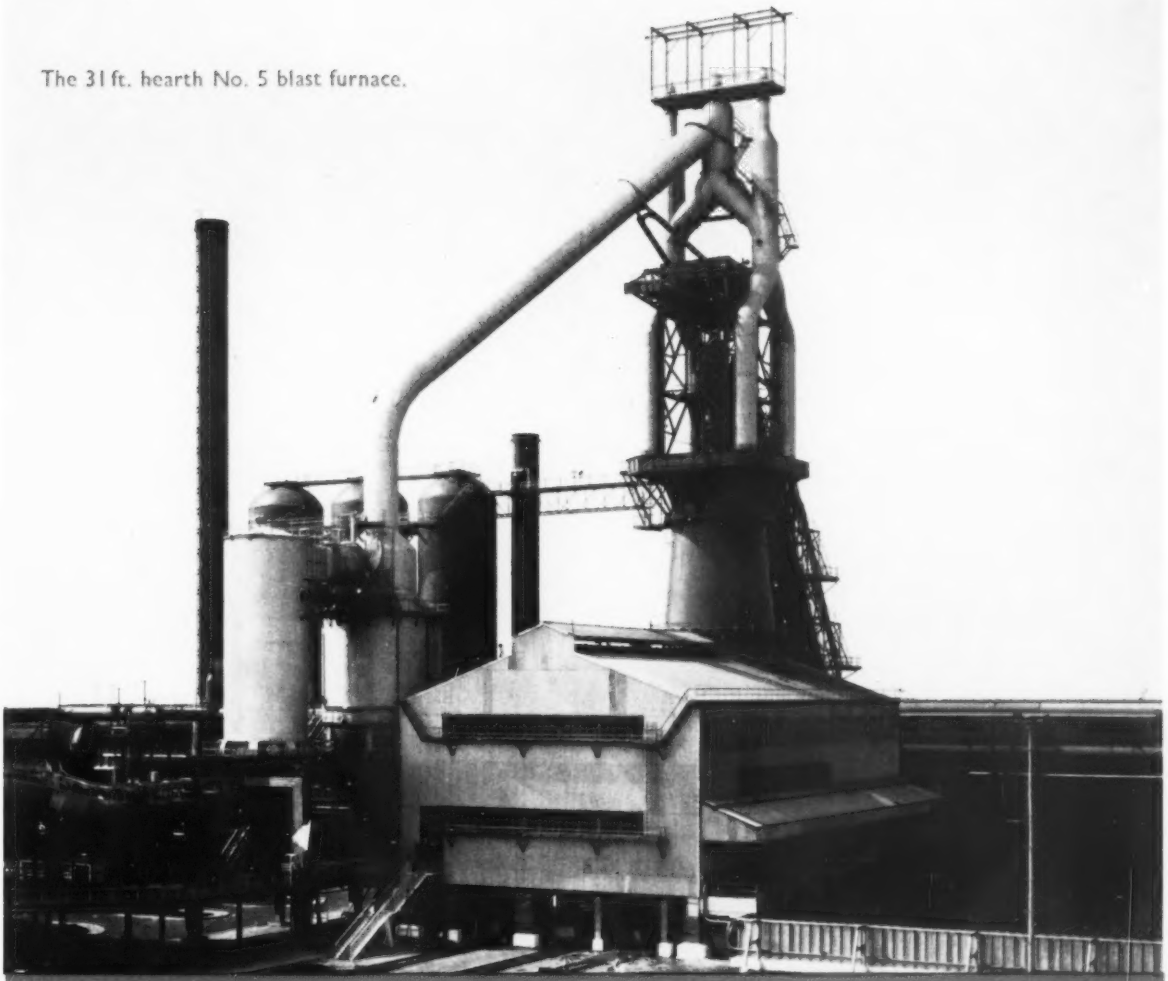

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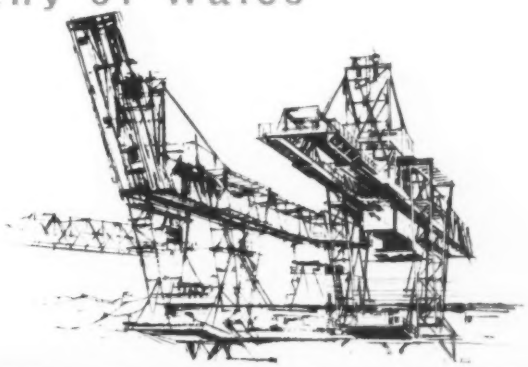

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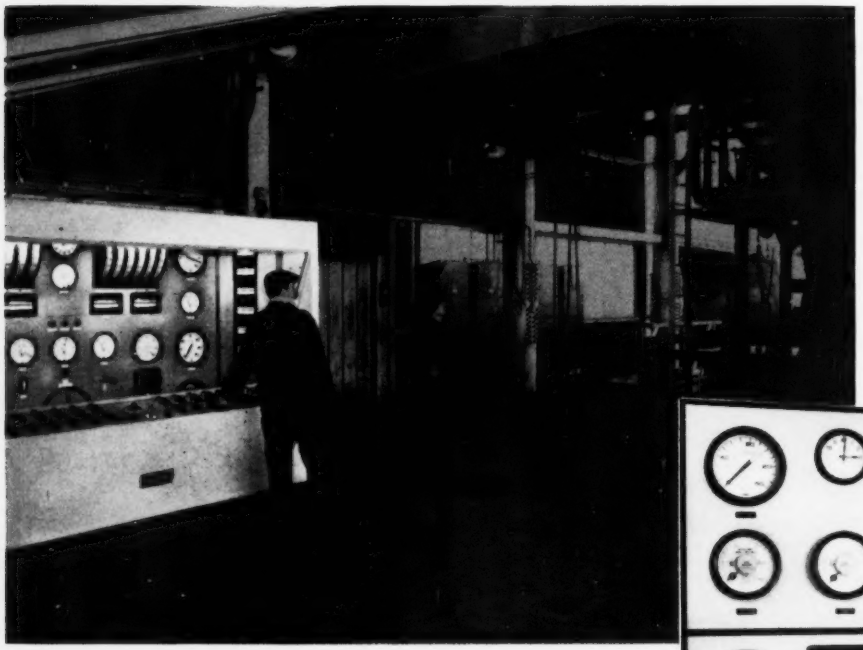
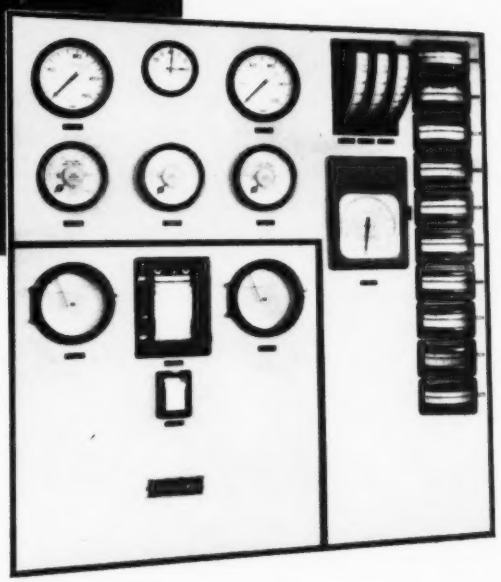


Recent installations in well-known U.K. steelworks.



"Toughness"— i.e., ability to resist adverse atmospheres, dust and vibration — is a "must" for this exacting duty.

Long-term, practical experience has given Electroflo instruments and controls the overall performance standards demanded by the iron and steel industries. Moreover, a comprehensive service is offered — the widest range of measuring and control systems, provision of complete instrument panels and control desks, installation and commissioning services by engineers with much steelworks experience. Further particulars on application.

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